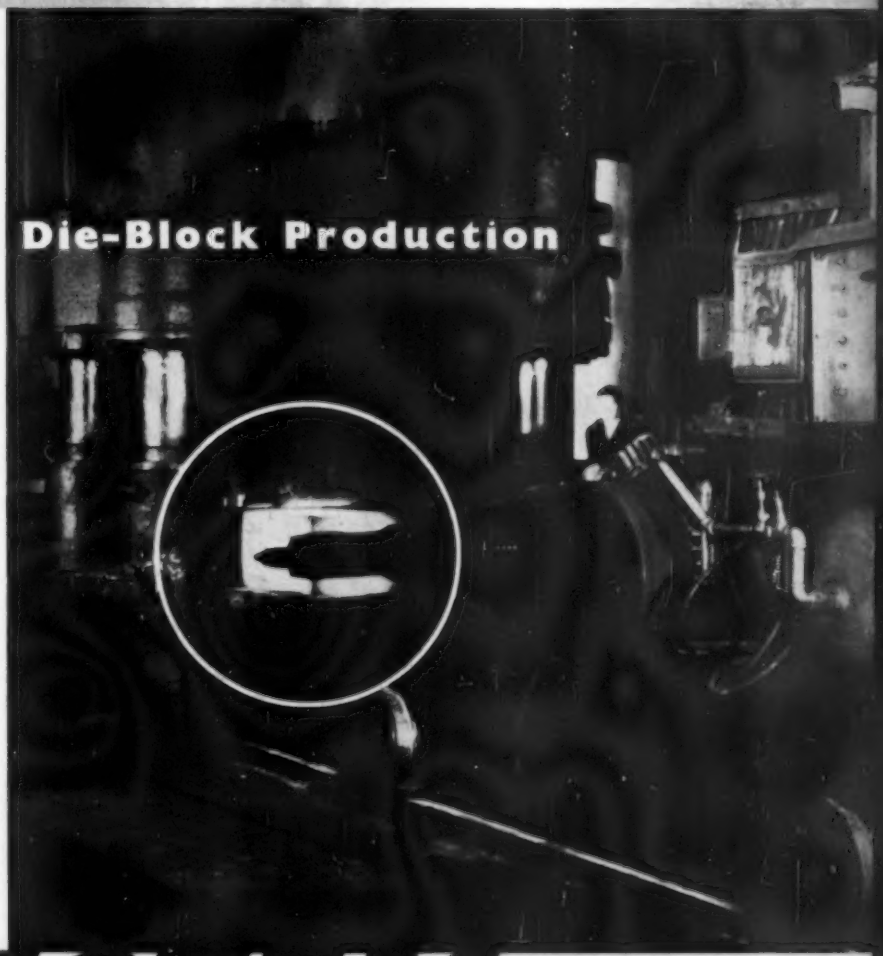
metal treatment

Vol. 26 : No. 166/7 JULY–AUGUST, 1959 Price 2/6

Specialised Die-Block Production

Electem die blocks are produced in the most up-to-date plant in Europe. Furnaces are automatically controlled; the Heat Treatment bay is designed to ensure absolute accuracy and uniform hardness and every block is individually checked. Our technical experts are always ready to call and discuss any die block problem.



SOMERS

ELECTEM

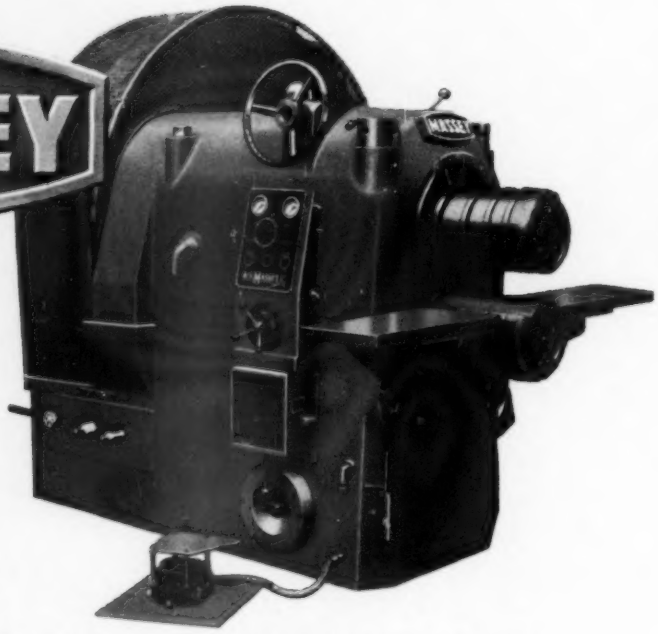
THE MODERN DIE BLOCKS

WALTER SOMERS LIMITED HALES OWEN NEAR BIRMINGHAM

SAVE MATERIALS MAN-HOURS

WITH

MASSEY



USE-ROLLERS

Rolling is now accepted as an excellent method, quick and cheap, of preparing material for drop forging, particularly in the case of long, thin, uses. Rolled uses cut out fullering, edging or forming and are clean, descaled and without flash. The material is accurately distributed, no operating skill is required, and time, labour and material are saved. The machine is portable, needing no permanent mounting or foundation and can therefore be placed just where it is wanted.

Massey designs include :—
Steam and Compressed Air Hammers, Pneumatic Power Hammers, Friction Drop Hammers, Double-acting Steam and Compressed Air Drop Hammers, Forging Presses, Trimming Presses, Tyre Fixing Rolls.

FEATURES OF THE MASSEY USE-ROLLERS

- Quicker and more positive roller adjustment
- Control and operation by foot pedal or automatic trip as required
- Well proved direct air operated friction-type clutch
- Automatic flood lubrication
- Central panel provides quick and easy control of motor, single or continuous action, and over-run adjustment

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MAKERS OF THE WORLD'S GREATEST RANGE OF FORGING PLANT

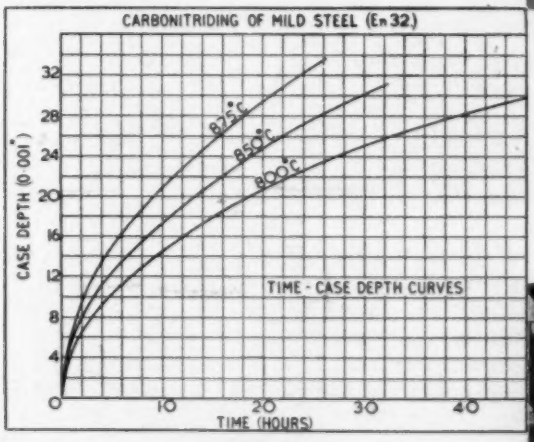
WILD BARFIELD CARBONITRIDING

. . . the best in every case

CONSIDER THESE POINTS:

- Clean working conditions
- Unskilled labour may be employed to carry out the process
- Post-cleaning operations reduced
- No storage space required for case hardening materials
- Simple system of atmosphere control

Shaker hearth furnace



- Atmosphere employed is raw Town's Gas and Ammonia
- No costly gas preparation plant required
- Batch or continuous equipment
- Consistent repetitive results

Rotary drum furnace

ELECTRIC WILD BARFIELD FURNACES

FOR *ALL* HEAT-TREATMENT PURPOSES

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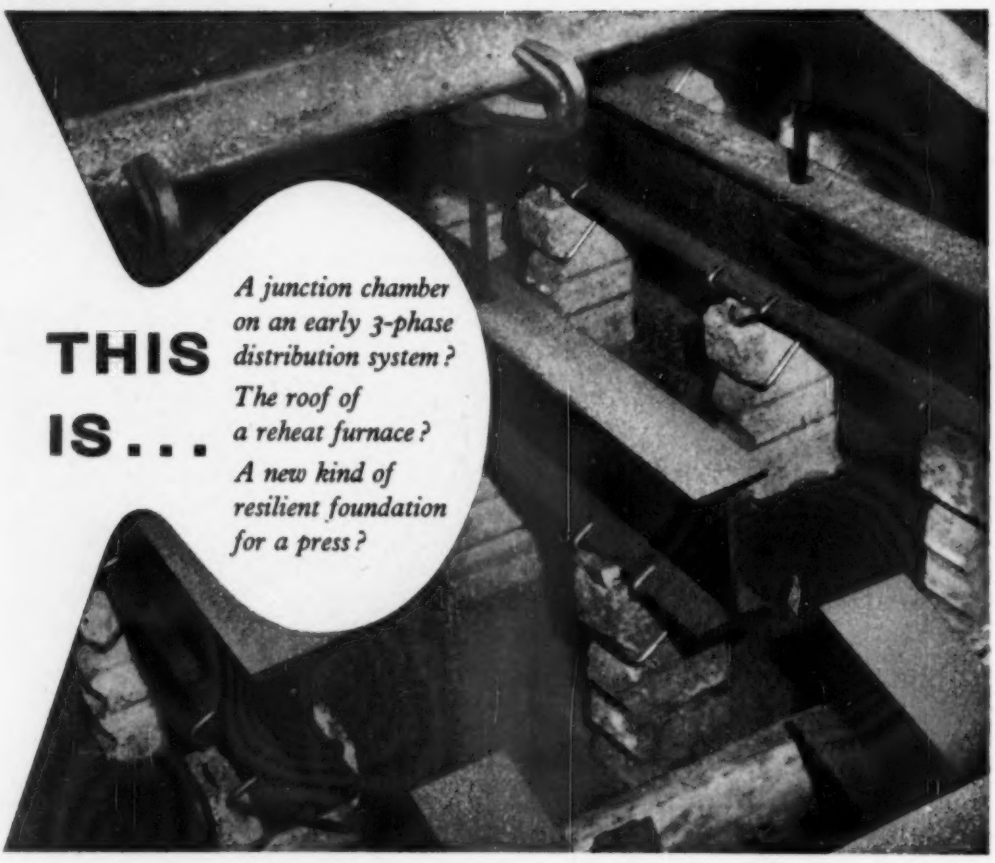
WB54


THIS IS . . .
A junction chamber on an early 3-phase distribution system?
The roof of a reheat furnace?
A new kind of resilient foundation for a press?

Number two is the right answer. It is part of the monolithic roof of a reheat furnace at the Margam works of the Steel Company of Wales, immediately before the refractory was placed.
The special anchor bricks hanging from the R.S.J's are secured by heat-resisting clips and will provide support through the full thickness of the refractory lining.
This roof was cast in "TRI-MOR" High Temperature Castable. It is a good example of the type of furnace lining which "TRI-MOR" refractories and Morgan's whole-hearted belief in monolithic structures have made possible.

Some of the most useful of the "TRI-MOR" refractories are:—

TRI-MOR Standard Castable
TRI-MOR High Strength Castable
TRI-MOR High Temperature Castable
TRI-MOR High Temperature Mouldable
TRI-MOR Dense 'Guncrete'
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NESTON, WIRRAL, CHESHIRE. TELEPHONE: NESTON 1406

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Fur by Bradleys, London

Do you really think I need bother about **low residuals?**

Some people do . . . . .

in carbon steels required for certain high duty applications.

Residual elements in Workington Electric Arc Furnace Carbon Steel are as follows

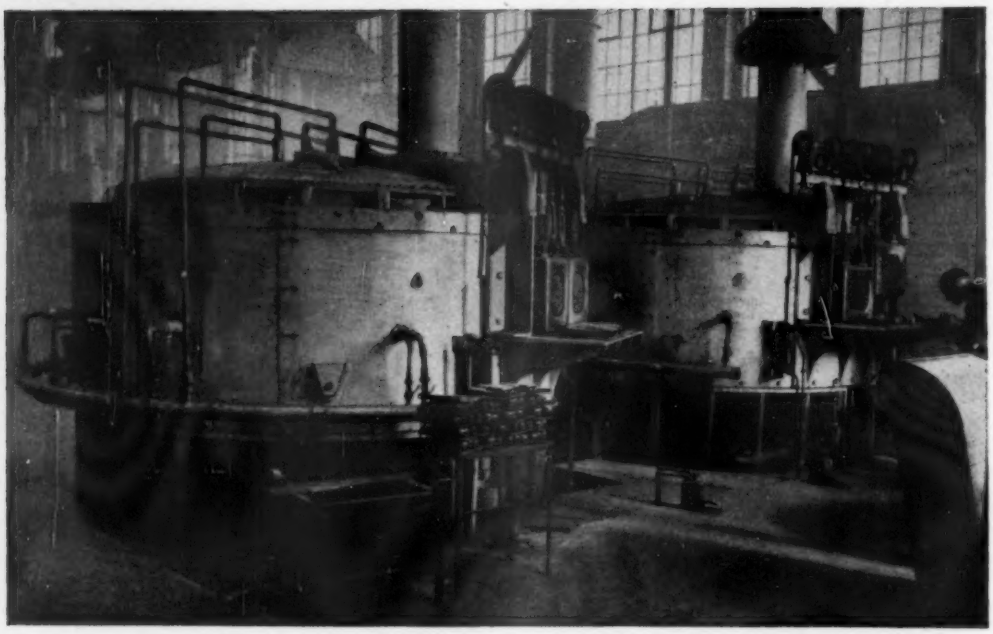
	average	maximum
Nickel	·01	·02
Chromium	·01	·015
Copper	·03	·04
Arsenic	·015	·02
Tin	·015	·02
Vanadium	·02	·04
Molybdenum	*below* ·01	
Lead	*below* ·01	
Cobalt	·005	·008
Antimony	*below* ·01	
Tungsten	*below* ·01	
Titanium	*below* ·01	

WORKINGTON

IRON AND STEEL COMPANY *Workington · Cumberland*

a branch of THE UNITED STEEL COMPANIES LTD

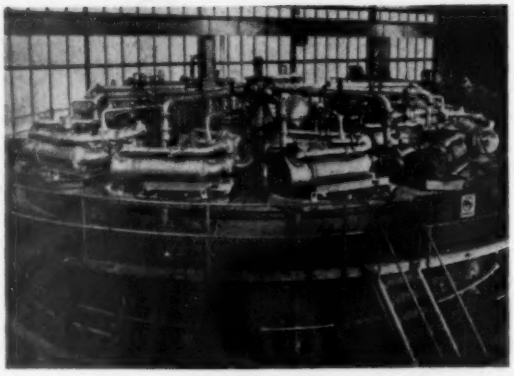
W 117



Back where they started . . .

An everyday expression which has more than a suggestion of fruitless effort, of time and energy wasted. Yet in production layout the converse is often true : time and manpower can frequently be saved by intelligent use of circular flow. The rotary furnace is an ideal embodiment of the principle, since the work is heated throughout its travel, and charging and discharging points are side by side.

And of course Incandescent rotaries have a further advantage, since they can be fitted with the Equiverse system of scale-free heating. The illustration on the right shows such a furnace with Equiverse firing, used for heating steel billets for extrusion at a rate of 10 tons an hour. Steering box components are being heat-treated in the two open-fired rotaries shown above.



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SIZES IN STOCK TO SUIT ALL HAMMERS

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EXTRACTOR FOR
RECLAIMING CUTTING
OILS AND COMPOUNDS
FROM SWARF

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by efficiently reclaiming cutting oils and compounds with BROADBENT CENTRIFUGAL OIL EXTRACTORS, and by leaving cleaner scrap which demands higher prices and reduces transport costs

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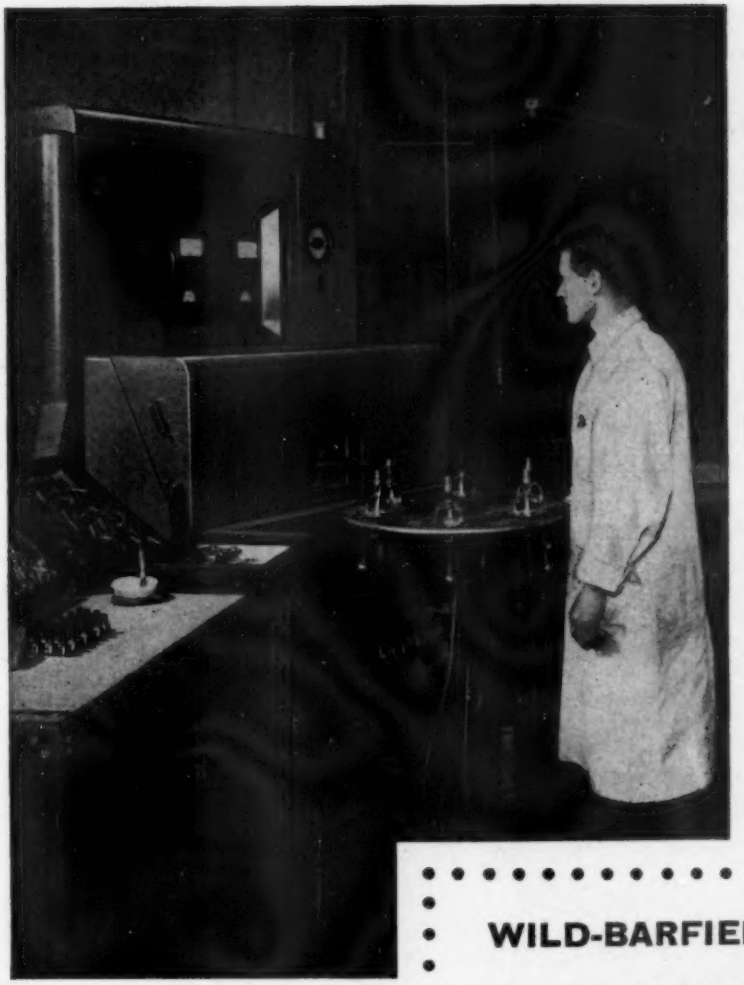

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WILD-BARFIELD A.H.F.

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WILD BARFIELD A.H.F.

Induction heating speeds production

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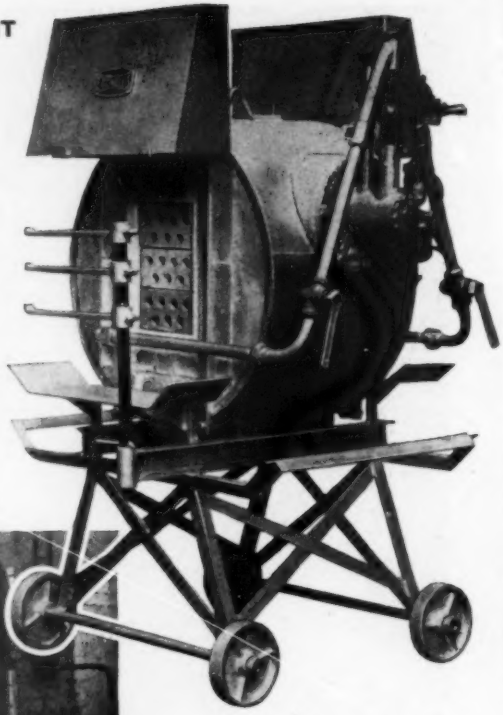
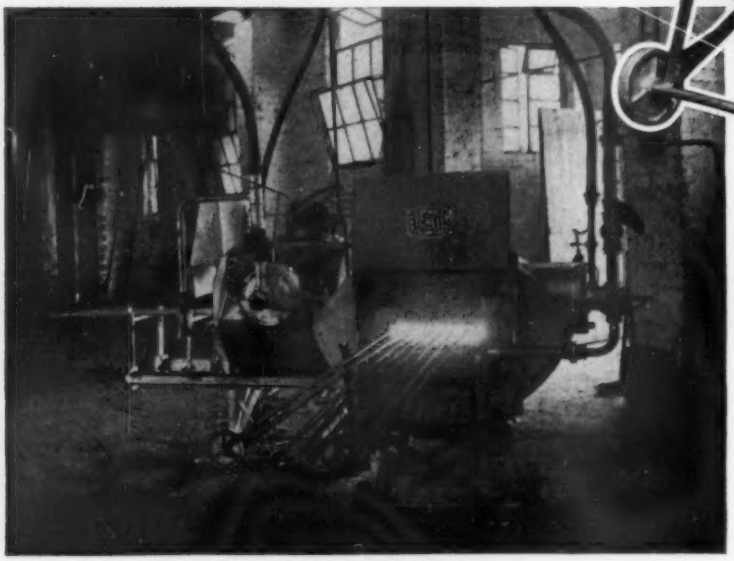


Oil-fired Bolt or Pin Forge Furnace



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IMPROVED OPERATING CONDITIONS.
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LESS WASTAGE, AND INCREASED OUTPUT.

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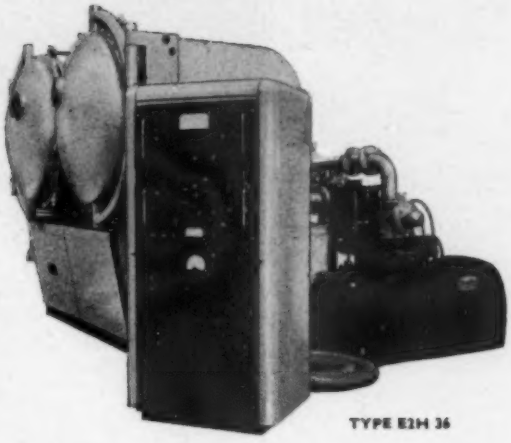

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THE FIRTH-DERIHON STAMPINGS LIMITED, SHEFFIELD

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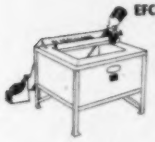

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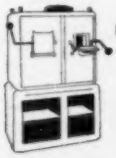
for austempering and martempering up to 500°C.
Publication R.18.



HIGH SPEED STEEL FURNACE—TWIN CHAMBER
Preheating up to 1000°C. and hardening up to 1350°C. without decarburisation.
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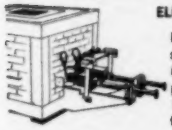
EFCO-UPTON CONTINUING GRAPHITE ELECTRODE SALT BATH



Hardening high speed steels, aluminium brazing and all processes up to 1300°C. with bath free from metallic oxides.
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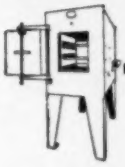

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for fast and accurate tempering and general heat treatment up to 750°C.
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For tempering, stress relieving, aluminium alloy heat treatment and all processes up to 700°C.
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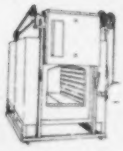

EFCO ELECTRIC SALT BATHS
Cyanide hardening, carburising, neutral treatment to 1000°C.
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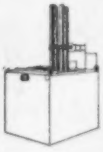

EFCO SALT BATH FURNACE MIDGET TYPE
with interchangeable pots for processes from 600 to 1350°C. for small outputs.
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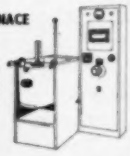

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Diamond Block method of atmosphere control for high speed steel hardening up to 1350°C. Vertical or horizontal.
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for heat treatment of long or slender parts.
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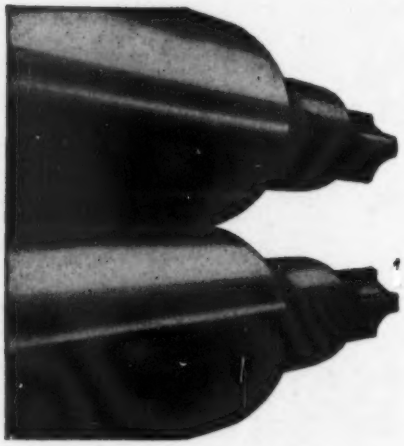

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HARDENED STEEL ROLLS

Made in the widest range of sizes to suit the requirements of every branch of cold rolling, from the small rolls for precision metals to rolls weighing several tons each. Doncasters rolls for a good job!

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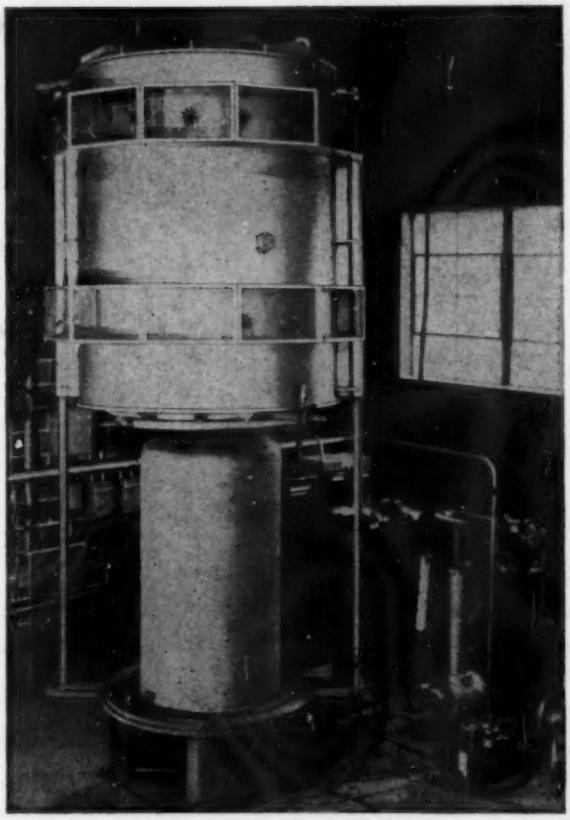
Vacuum Heat-Treatment



This model has a uniform hot zone 36" dia x 36" high and ultimate vacuum of less than 1 micron.

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The experience of the National Research Corporation who have built more than 100 vacuum furnaces, now in successful operation, and the facilities in Britain of Wild-Barfield combine to offer unsurpassed vacuum resistance furnaces of the horizontal muffle, pit and bell types with diameters from 2 to 54 inches.

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- **Induction and Arc furnaces**
- **Hydrogen in Titanium analyser**
- **Vacuum Fusion gas analyser**
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* NRC *is the trademark of the National Research Corporation, registered in the United States Patent Office.*

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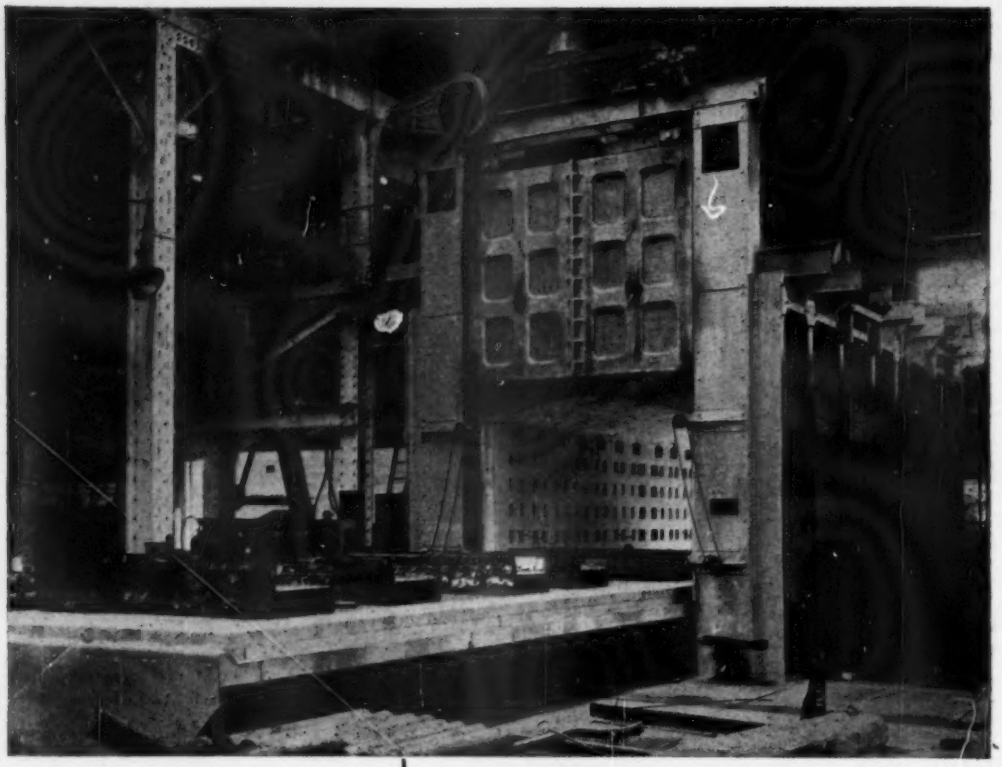

Photograph by courtesy of Messrs. T. Firth & J. Brown Ltd., Sheffield

Brayshaw

Town's Gas Fired Bogie Hearth Furnace

of the products recirculation type tempering Furnace installed at Messrs. T. Firth & J. Brown Ltd., Sheffield

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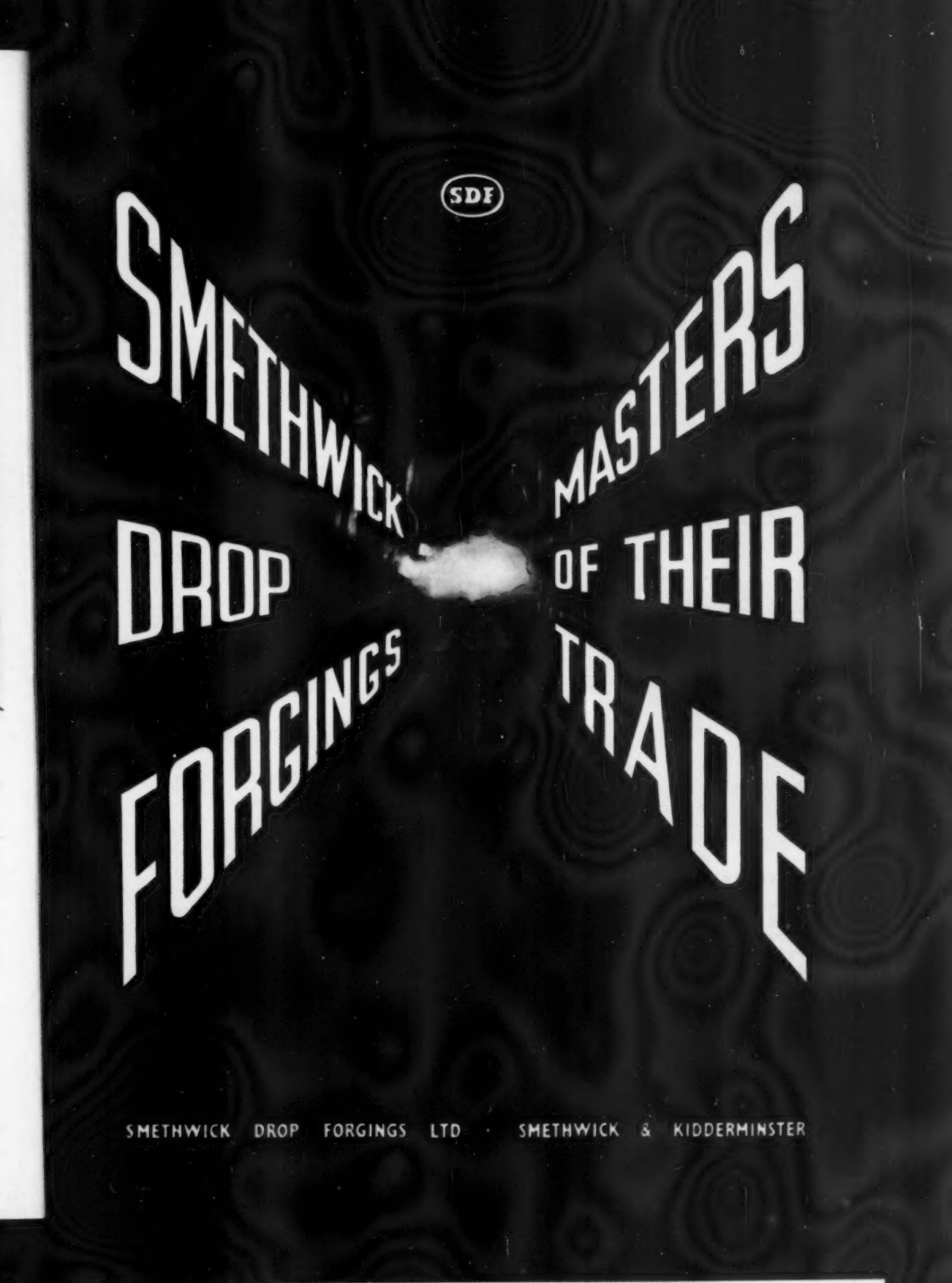
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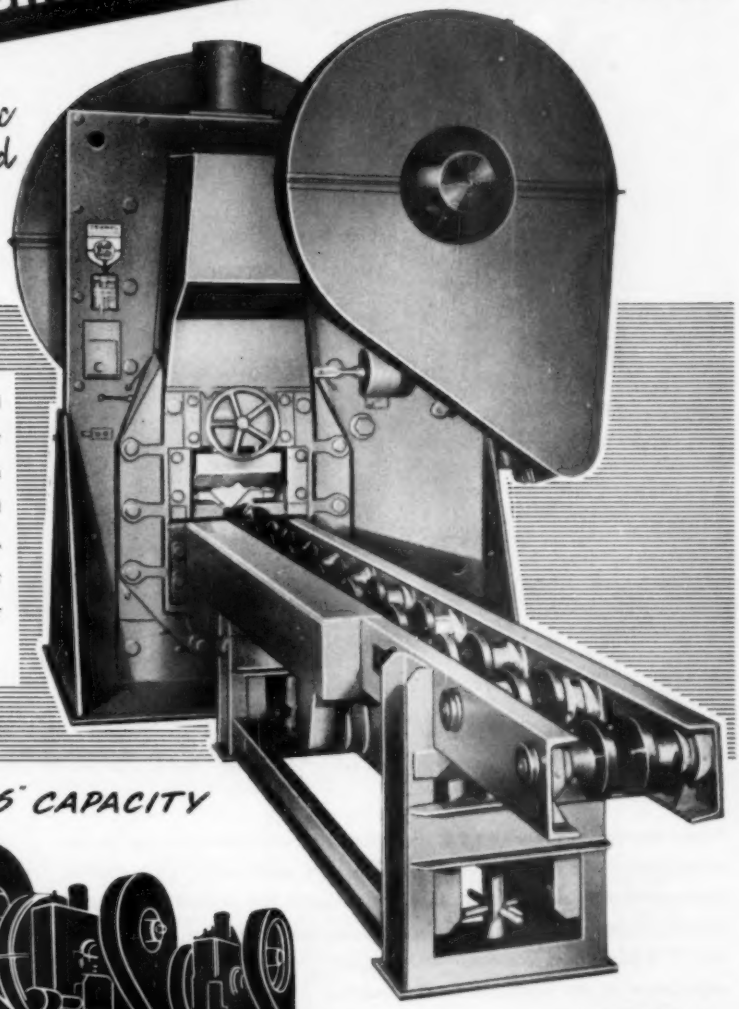

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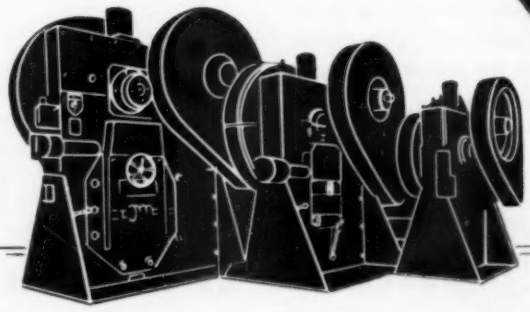

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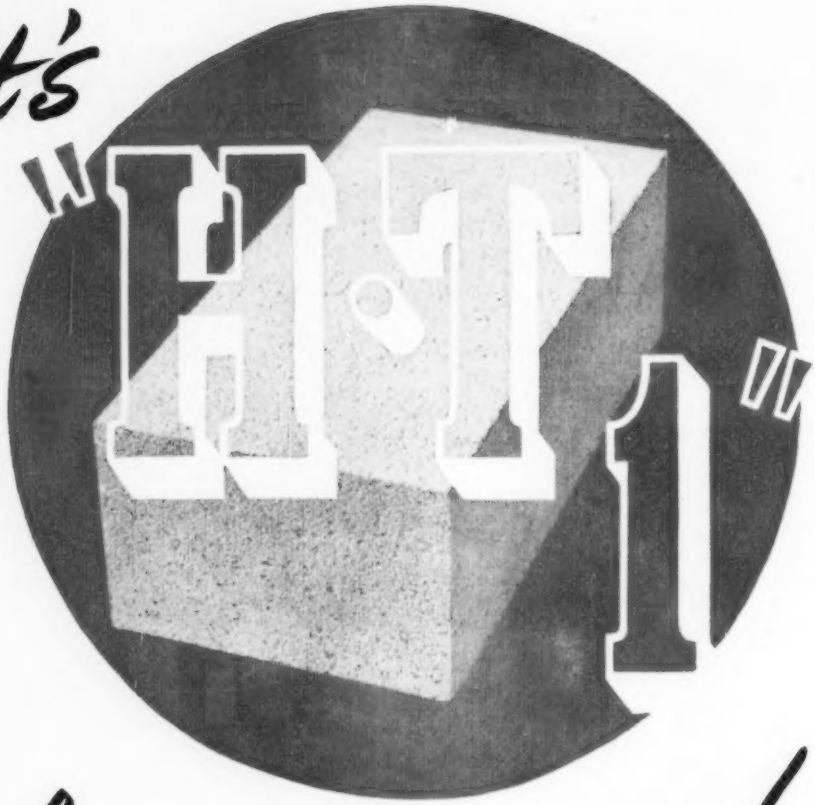

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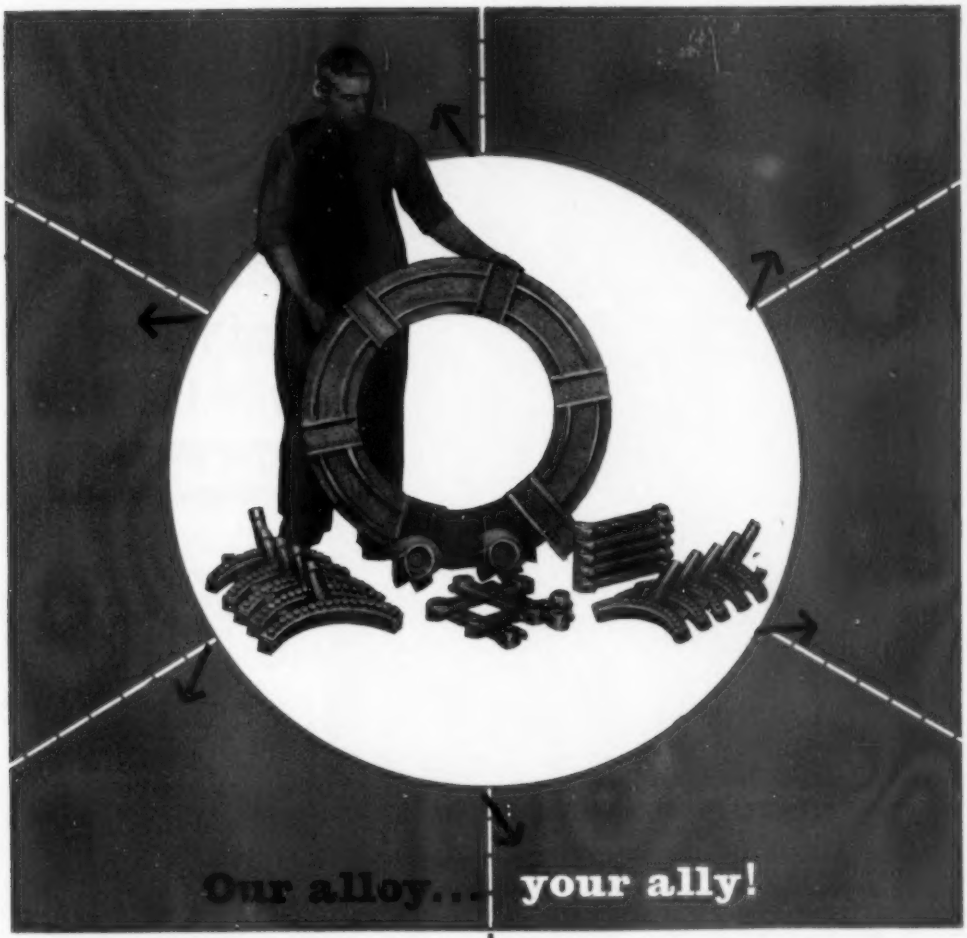

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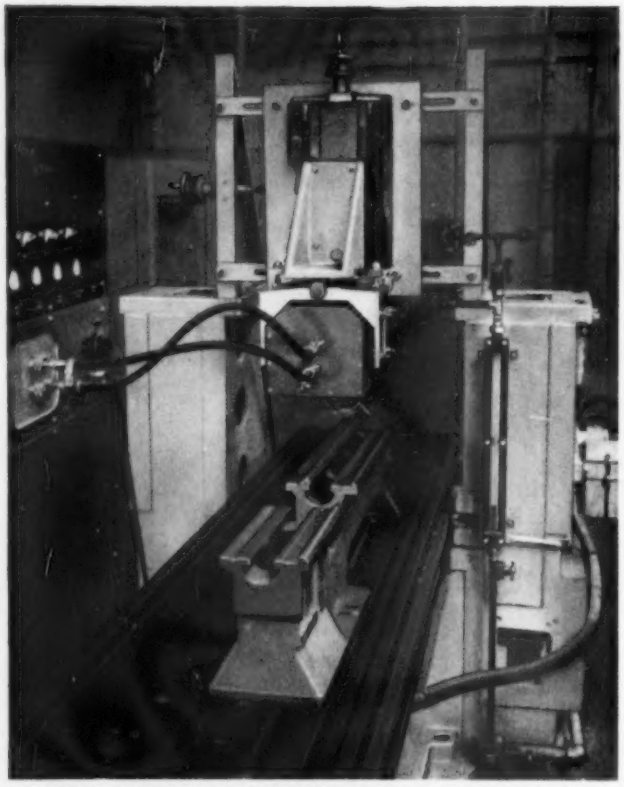

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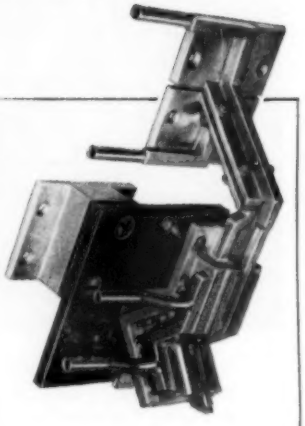

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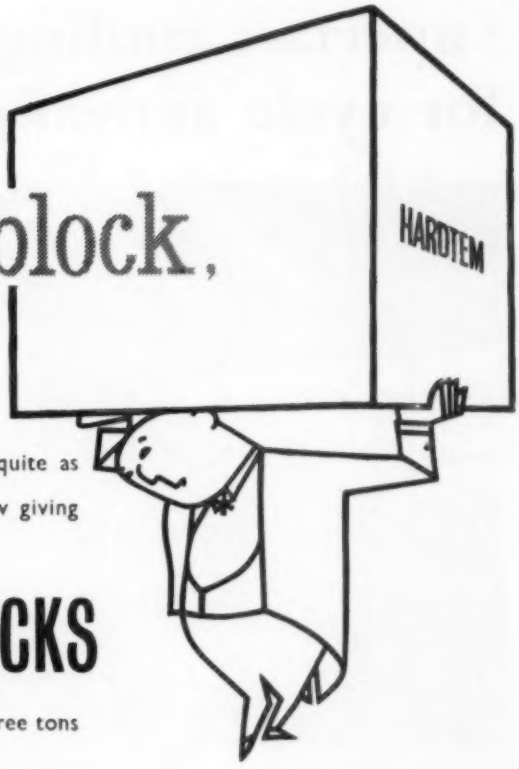

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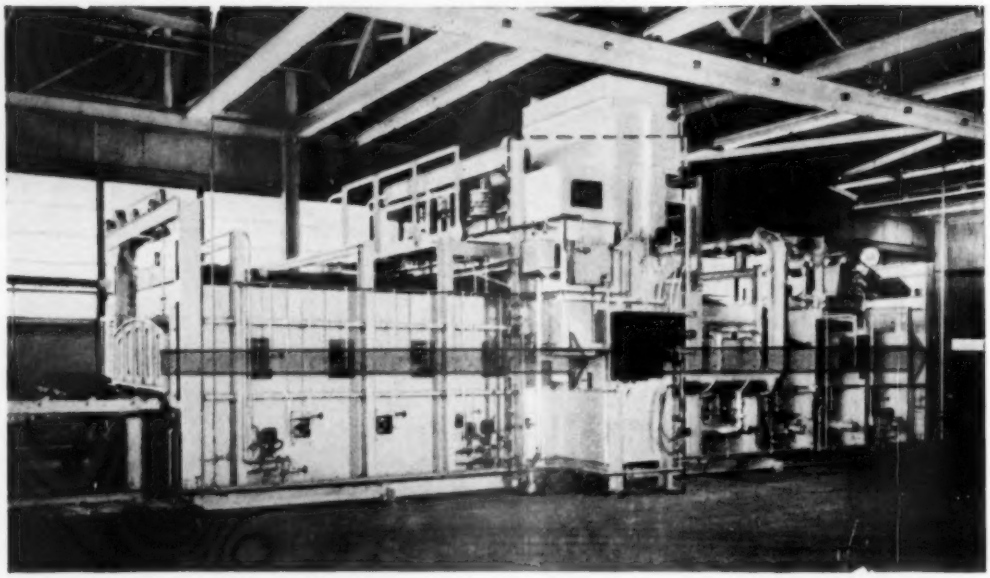

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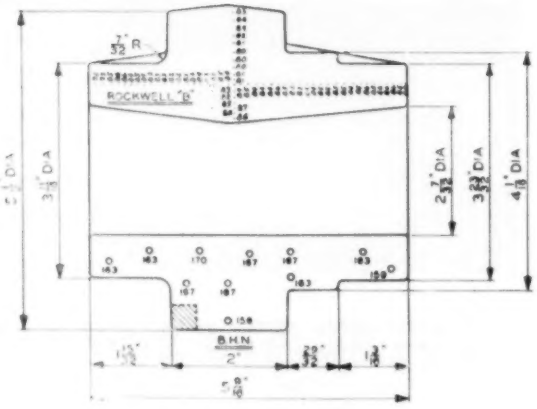

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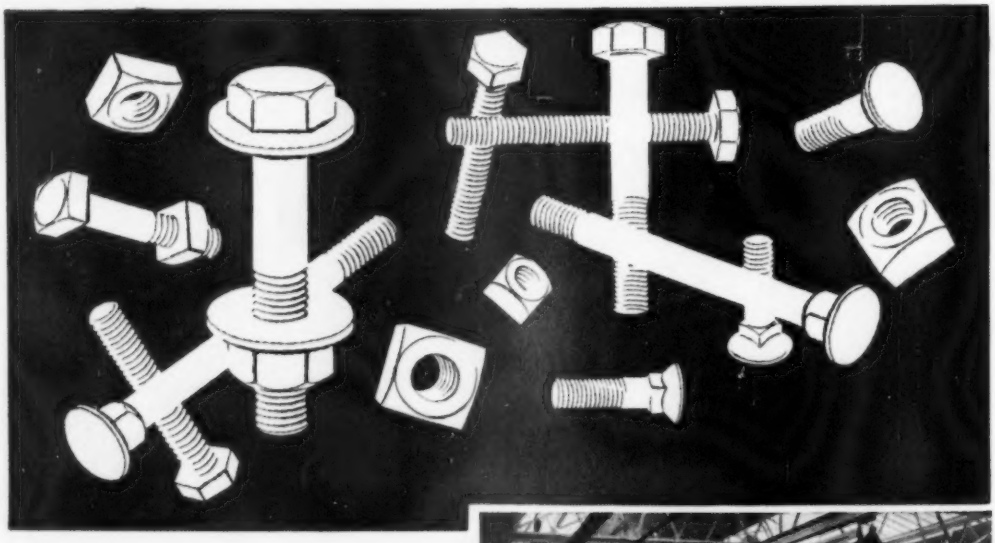

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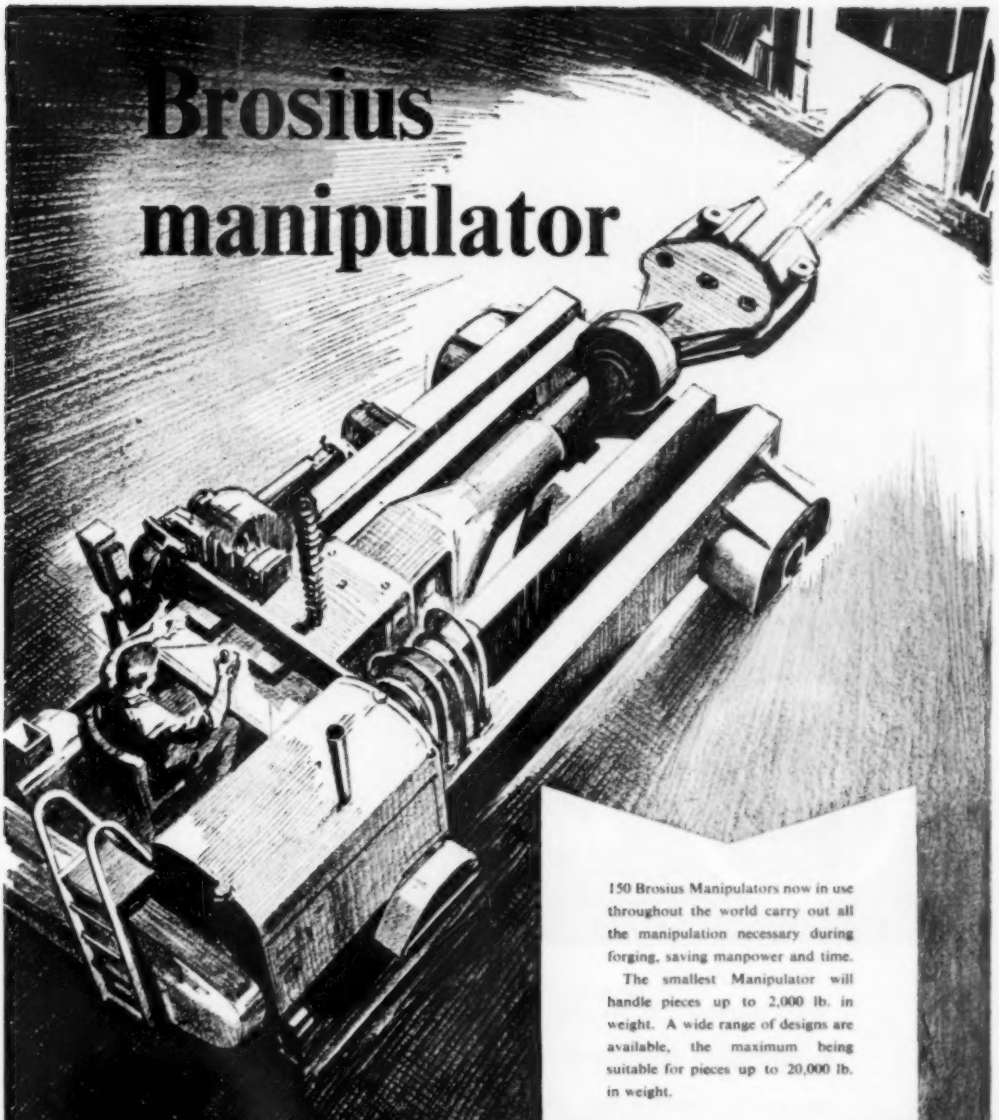

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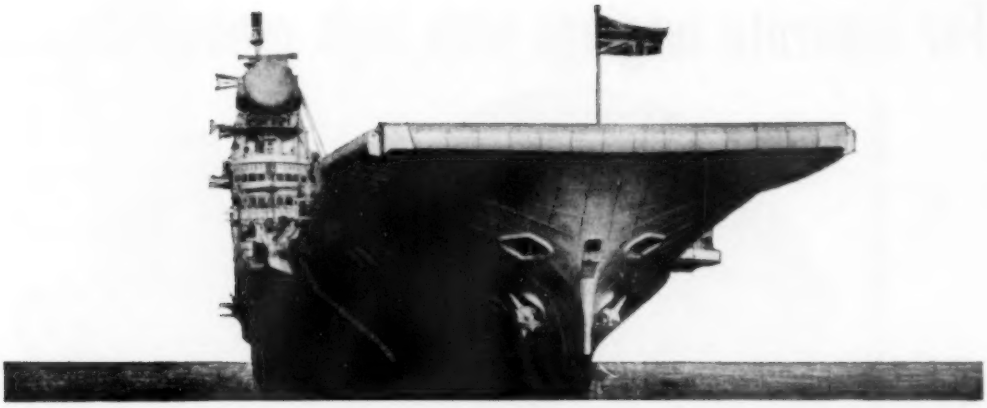

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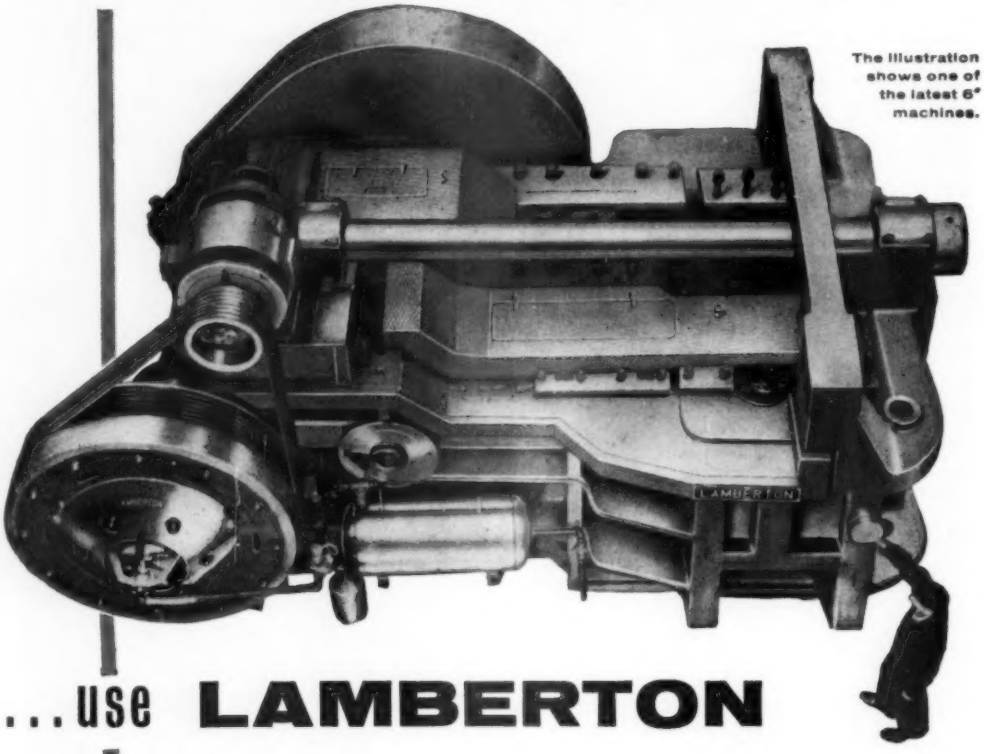

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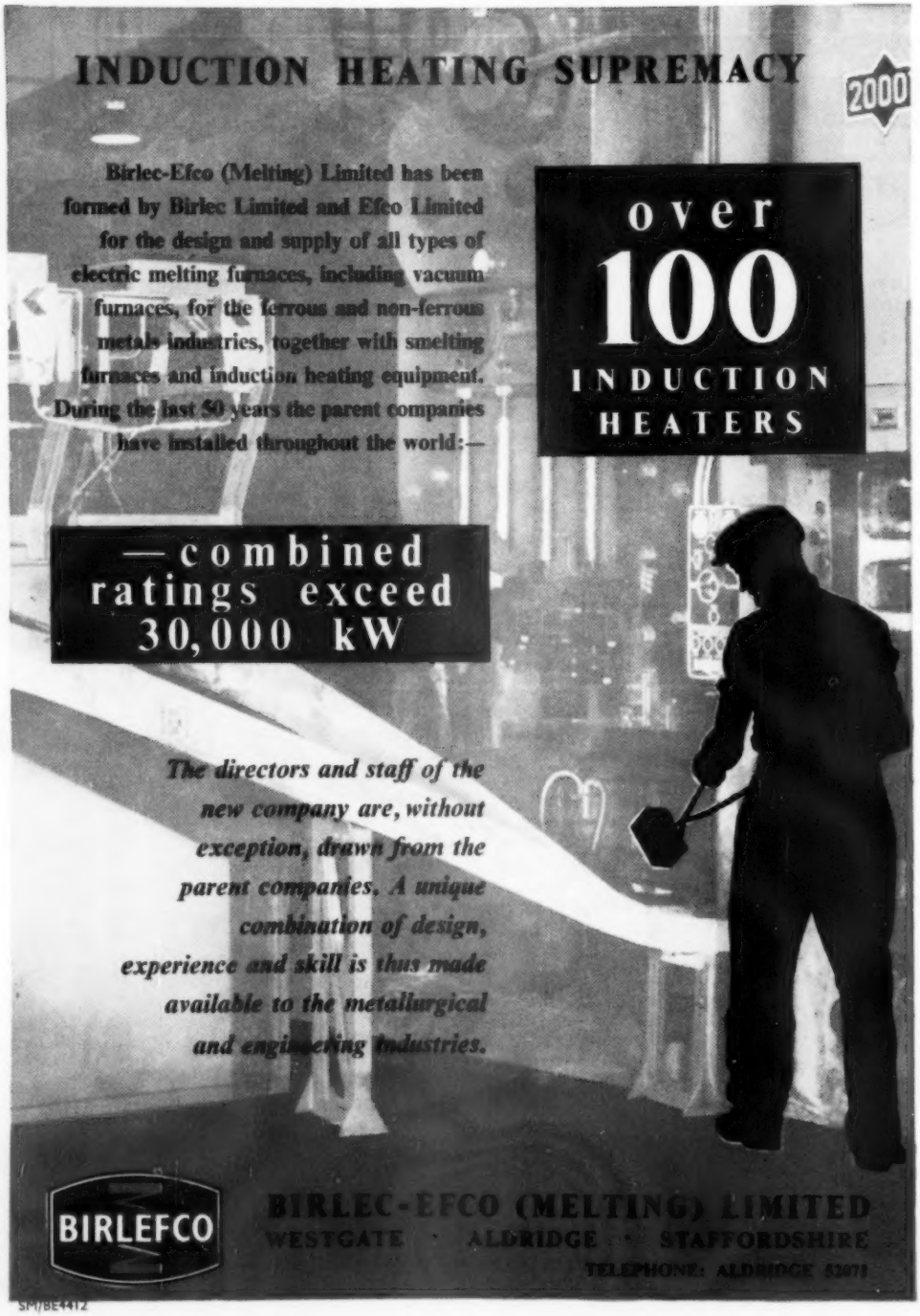
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July-August 1959

Vol 26, No 166-7

Metal treatment

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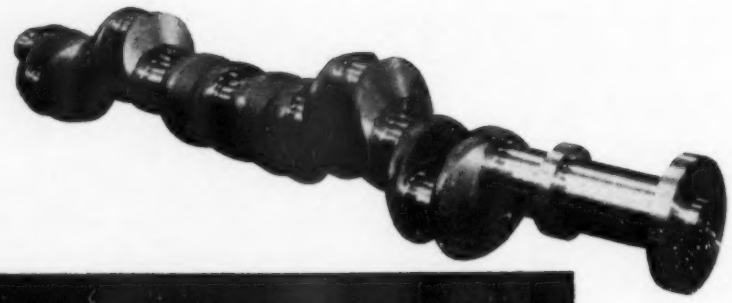
We very much regret that a separate publication for July METAL TREATMENT was prevented last month by the printing dispute. Work has now been resumed in the printing trade, so we are presenting a combined July-August issue, and hope that the September issue will appear as normal. Annual subscribers will, of course, have their subscriptions extended to cover the lost issue.



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A national asset?

NEXT month, the S.B.A.C. annual flying display and exhibition will once again take place at Farnborough, and its advent acts as a reminder of the close connection between developments in aircraft design and trends in general engineering techniques. In view of the present concern felt by the British aircraft industry in respect of its future, it may be as well to examine the relationship between aircraft design and metallurgical development, and attempt to foresee possible repercussions which might result from a curtailment of the activities of the aircraft industry in this country.

That some, at least, consider there to be a serious outlook for the future of British aviation may be judged from the following remarks quoted by Mr. Peter G. Masefield, managing director, Bristol Aircraft Ltd.: " Unless a clear and concise Government policy towards research, development and procurement of British transport aircraft is formed within a very short while, there will be no further major British transport aircraft beyond the generation now going through its labour pains."

Mr. Masefield was speaking at the conference on the theme ' The aircraft industry—a national asset,' held in April this year at the University of Southampton and promoted by the Institution of Production Engineers. At the same conference, Mr. S. P. Woodley, M.B.E., director, Vickers-Armstrongs (Aircraft) Ltd., in his talk on the impact of the aircraft industry on other industries, gave many examples where problems of aircraft design may be claimed to have led to developments in the metal industries.

The first examples cited by Mr. Woodley are naturally the ' new ' metals and alloys. After mentioning that the use of aluminium in civil and engineering structures had been directly stimulated by the initial search for lighter and stronger aircraft structures, he goes on to say that the development of high-strength light alloy materials, and high-strength steels, was almost entirely due to metallurgical researches initiated by the industry. If titanium was ever applied commercially, or in the general engineering sphere, it would be entirely as the result of development in satisfaction of the aircraft industry. Mr. Woodley then goes on to deal in greater detail with recent developments in metal working which have resulted from research prompted by problems in aircraft construction.

' It is in the development of machine tools and adaptation that the impact of the aircraft industry has been most marked. Until quite recently, the amount of machining necessary in the manufacture of airframes was comparatively small, but new design philosophies involving the integration of components has expanded the machining processes required until they now occupy a premier position in aircraft construction.

'At first, there were almost no machine tools available which could produce the required complicated sculptured components rapidly and economically. The only machines which could undertake the task to the accuracies required were expensive copy milling machines, but within a short time new machine tools, employing high-speed milling heads and copy-milling techniques, were devised to meet this relatively new and complex requirement. This in turn has culminated in the growth of a separate industry, providing machinery that has revolutionized the production of piece-parts in light alloy materials.

' A new method of metal removal has recently been developed by the aircraft industry which will, without doubt, find many uses in other industries. It is now being investigated as a means of obtaining motor body panels with a weight relief by pocketing methods. This is the principle of machining by chemical erosion. The idea of metal removal by masking the material where it is not required to be removed, and then

immersing it in a chemical bath for a period of time dependent upon the depth of etch required, is so simple and economic of labour that it must have a wide future appeal in other industries.

'In the manufacture of jet engine components new problems were posed, and specialized requirements arose in connection with the production of turbine and compressor blades and in the manufacture of flame tubes and turbine chambers; requirements that had to be met by the development and adaptation of existing types of machining tools or by invention. Arising from this need have come a large family of machine tools for the generation or straight machining of turbine blades to difficult aerodynamic forms to extremely close tolerances.

'The industry uses large numbers of sheet metal components which were at one time produced variously by press and hand methods, but the introduction of processes designed to increase output led to the development of the stretcher press, a type of forming that is now being used in industry, as an economical and ready means of manufacturing difficult double-curvature sheet metal components without the necessity for expensive double-acting press tools. The rubber die press, too, is a unique aircraft industry innovation, which has involved large-scale development and tooling methods specially devised for the process which now virtually eliminates post-dressing of component parts. Later improvements of the stretcher technique have developed the stretch wrap process, the Verson Wheelon and the hydro-form presses, all of them peculiar to the requirements of the aircraft industry, but which have revolutionized the production of sheet metal components to the ultimate advantage of other industries.'

When we consider Mr. Woodley's claims, and remember that many aircraft companies in the U.K. are being forced to switch their activities to other specialized forms of production, we may feel that quite apart from the larger question of overseas commercial aircraft competition, the metallurgist has a very direct interest in the future of the British aircraft industry.

Management training for steel industry

The establishment this summer of the British Iron and Steel Federation Management College at Ashorne Hill, near Leamington Spa, is the culmination of several years' experimental work by the Federation in organizing management courses, first on a regional and then on a national basis.

The creation of the college, to continue and develop the work already done, underlines the industry's recognition of the need for the highest possible quality of management. Enquiries recently made of the 40 companies which have sent 300 men from 70 works on the courses have strongly confirmed their value.

The courses now running at Ashorne Hill were originally proposed in a report on education and training for management in the industry approved by the Federation Executive Committee.

The courses are of two types. The first, intended primarily for larger companies, is of eight weeks' duration, split into two halves separated by 10 weeks. The second, consisting of two weeks followed at a later date by two separate weeks, is designed particularly to meet the requirements of men from small companies or from small units in large groups or companies.

Hot working of alloy steels

Part 1: experimental methods

A. GUEUSSIER and R. CASTRO

*This investigation into the hot ductile behaviour of alloy steels of varying compositions under varying conditions was described by the authors at the 1957 autumn meeting of the Société Française de Métallurgie, Paris, and published in French in 'Revue de Métallurgie,' 1958, **15**, (11). The authors are both with Aciéries Electriques d'Ugine, Savoie, France, Mr. Castro being chief engineer, head of research*

IN THE HOT WORKING of alloy steels, there are a number of different forming tools at our practical disposal, of which the most important are the hammer, the forging press, the rolling mill and the wire-drawing press. Actually the reduction of the steel from ingot to bar or sheet is often done in several heats, with differing tools from one heat to the next. But each tool works under specific conditions, which can be summarized as follows:

1. Reduction in the one heat can be done either in several passes, as is customary with the drop hammer, the forging press, bar and wire rolling mills, etc., or in one single pass, as is the case with the wire-drawing press, the inclined tube mills and certain die-stamping operations. When several passes are used in reduction, the temperature of the metal falls throughout the whole process of working.

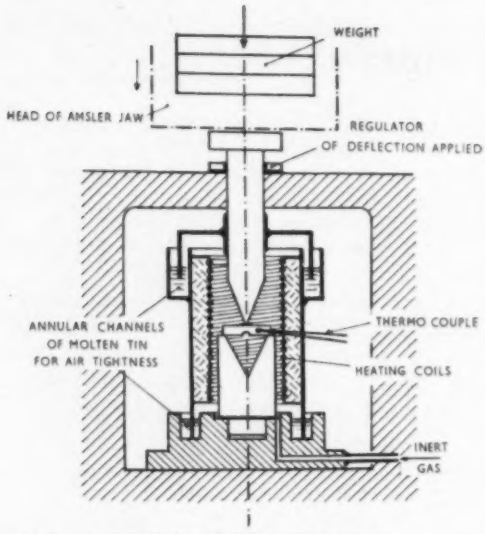
2. Working is done with very different average speeds according to the kind of tools used. To give a comparative idea, let us say that the drop hammer can reduce the steel at a speed of the order of 50,000%/sec., the extrusion press at a speed of 5,000%/sec., the ordinary bar mill at a speed of 500%/sec., and a large forging press at a speed of 50%/sec. These figures vary according to the tools used and to the sizes of the work-pieces involved, and they are only given here to establish some order of magnitude. They do show, however, the variations in speeds of reduction realized in practice in the work shops.

3. The strains set up in the steel can be very different; for example, in tube making, a push-bench mill works the metal chiefly in the longitudinal direction, whereas obliquely inclined mills, of the Mannesmann or Stiefel type, impose stresses on the rough work-pieces in the transverse direction.

The essential problem before the operator is to fix for each type of alloy steel the operating conditions to be applied so as to ensure full hot working of the steel without cracking. This problem involves the fixing of exact conditions for the heating of the steel, the limits of temperature permissible in rolling, the ratio of reduction allowable in each heat, and the realization of the programme required. Settling these conditions industrially for the working of any given alloy steel is a long and difficult matter and involves large quantities of metal, if it is carried out directly in the shop. It is advantageous to carry out preliminary investigations in the laboratory so as to fix the essential factors which will need to be checked during working. In order to solve this complex problem it is possible to have recourse to various experimental methods. Described below are the methods which have been used in this investigation, pointing out the advantages and disadvantages, and different examples are given of the results obtained. This investigation will show that the behaviour of hot steel depends largely on the conditions of working but also on its structure and composition as well as on steel-making conditions. It will thus be evident that the problems of working the steel can only be solved by co-operation with the metallurgist and the steelmaker.

EXPERIMENTAL METHODS

The first care of the worker of steel is to avoid cracks. Cracking corresponds to a local rupture of the steel, and it can be considered that there is



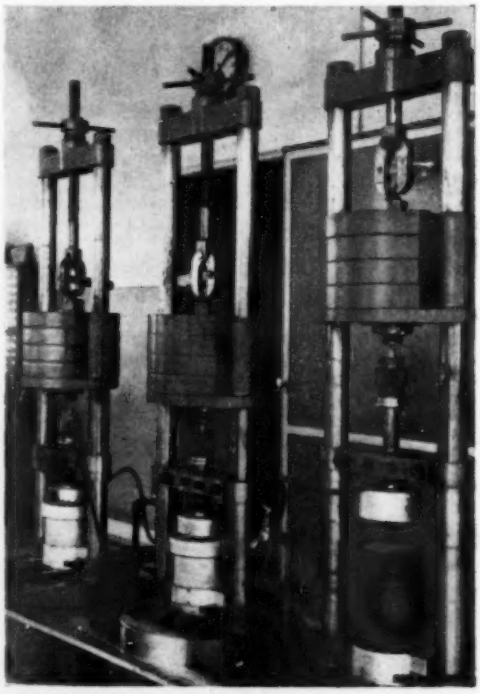
1 *Diagram of bend forgeability testing machine*

more risk of this happening in proportion as the actual ductility of the steel is less. The experimental analysis of working conditions therefore essentially leads to the study of ductility, when hot, of the alloy steels as a function of methods of working on the one hand, and of the analysis and structure of the steel on the other.[1] The authors have, therefore, endeavoured to carry out tests for deforming alloy steels to breaking point in conditions comparable with those found in industrial practice, and so determine the corresponding ductility of the steel. In these tests they have principally sought to vary the temperature and the speed of deformation and subsequently the method of working, either continuous or with several passes.

Bend tests

The bend test is one of the oldest tests suitable for evaluating the aptitude of a metal for hot working. It has for a long time consisted of bending through 180°, a flat test-piece heated to the required temperature and observing on the smallest radius of the bent portion whether cracks have appeared.[2] Such a test is, however, only a qualitative one. For the present investigation, the technique described by Portevin, Prétet and de Lacombe has been used.[1] The test-piece has either been the usual Mesnager impact test-piece or a small notched bar, 6 × 10 × 30 mm. ($\frac{1}{4}$ × $\frac{3}{8}$ × $1\frac{3}{16}$ in.). The notch cut in the test-piece is only put in to localize cracks for easier observation. The test-piece is subjected to bending in a Kanthal coil furnace, with direct radiation without muffle, which can be taken up to 1,300°C. The working part of the furnace is insulated from the outside air and a current of nitrogen or argon is swept through it. The lower V support, on which the test-piece is placed, as well as the knife-edge which causes bending, is kept in contact with the test-piece in the furnace during the whole of the test (fig. 1). The test is based on the observation of cracks corresponding to deformation to a given deflection. The furnace and its accessories are placed either on special hammers (fig. 2), which allows of working at a deformation speed of about 20,000%/sec., or on an Amsler tensile test machine, which can be worked at much slower speeds of deformation varying between 1 and 30%/sec. In the first case, the movement of the knife-edge is governed by that of a 60 kg. (132 lb.) weight falling free from a given height (fig. 2). In the second case, movement of the knife-edge is directly controlled by the adjustable head of the Amsler testing machine. In both cases movement of the knife-edge during the test is limited by a system of adjustable stops placed outside the furnace, which fixes the amount of deformation applied to the steel.

The advantages of this method are: (*a*) Reduced



2 *Hot shock bend test machines*

size of test-pieces, which is a help in obtaining them, particularly from steel as cast. (*b*) Ease of experimenting when working in an inert atmosphere, and also of studying the effect of different atmospheres on the test results. (*c*) Possibility of making large variations in speed of deformation. (*d*) Possibility of carrying out, easily, deformation by successive passes and heats.

Quantitative results can be obtained from this bend test by determining for each test condition the critical deflection, f_c, corresponding to the appearance of the first crack, but this method calls for a large number of test-pieces. Moreover, this test is only suitable for fairly fragile materials, deformation being limited to a maximum deflection governed by the geometrical conditions under which the test is carried out.

Torsion tests

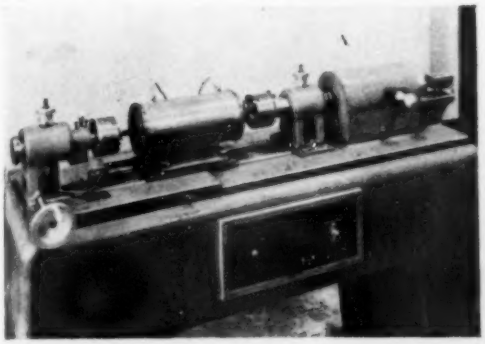
Dr. A. Sauveur was the first to use this kind of test. Machines using this kind of deformation have been described by various authors[3–6] and fig. 3 shows the torsion test machine developed in the authors' laboratories, based on various published ideas. The test-piece is heated in a furnace, without muffle, with direct radiation winding, which can get up to 1,300°C. It is held at its two ends in four-jaw chucks, cooled by air (fig. 4). The revolving end can be driven by an asynchronous motor with four-speed gearbox. The number of revolutions before fracture is registered by a revolution counter which is only connected to the driving shaft whilst the steel is being deformed. This connection is produced by the play of an electric clutch, of which the circuit is controlled by contact between a dog and its end stop. When the test-piece breaks, the dog resumes its vertical position under the influence of its own weight and by moving away from the stop opens the circuit in the electro-magnet.

The test-piece used consists of a calibrated portion, 8 mm. diameter ($\frac{5}{16}$ in.) × 50 mm. (2 in.) long, turned in the centre of a forged bar, 12 mm. square ($\frac{1}{2}$ in.) × 330–350 mm. long (13–14 in.). The average superficial speed of deformation applied to the steel in this test can vary between 12 and 100%/sec., according to the speed of rotation applied. Below the surface of the test-piece, deformation speed obviously varies with the distance of the given point from the axis. Deformation speeds are, therefore, heterogeneous through the mass of the test-piece.

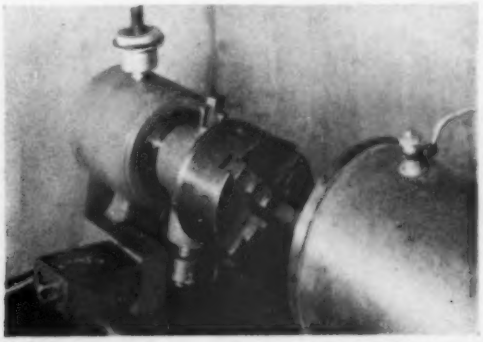
The testing machine has been developed in such a way as to make it possible to carry out torsion tests with test-pieces of fixed lengths, or of free lengths, this depending on whether movement is allowed or not in a longitudinal direction to the chuck holding the non-rotating end of the test-piece, in the corresponding bearing (fig. 4). In the standard machines described in previous publications, it would appear that the torsion test has always been carried out with fixed length test-pieces.

Tests have shown that austenitic steels, of the 18/8 type, when subjected to torsion tests with free length, give rise to a swelling in the steel, that is to say, the test-piece gets shorter and increases in diameter (fig. 5). The said swelling can be serious enough to force the calibrated portion on to a neighbouring section of the original square bar, thus causing a parasitic deformation of the latter. In tests of this kind, the steel is subjected to simple shearing stresses but fracture often occurs at the set-downs owing to the above-mentioned swelling, and this makes the results obtained of doubtful value.

If the length of the test-piece is kept constant



3 *Hot-torsion test machine. On the left is the semi-fixed jaw for holding test-piece and on right of the furnace is the movable jaw*



4 *Hot-torsion test machine showing the semi-fixed jaw and declutching dog*

during the test, a normal fracture is obtained from the same steels, this being generally sited about the middle of the calibrated section and the results are quite reproducible. In this case, the stresses set up in the steel during deformation are complex and are the result of the composition of simple shearing stresses, as above mentioned, with a draw stress along the axis, which reflects the opposition of the mechanical fitting to the natural swelling tendency of the steel.

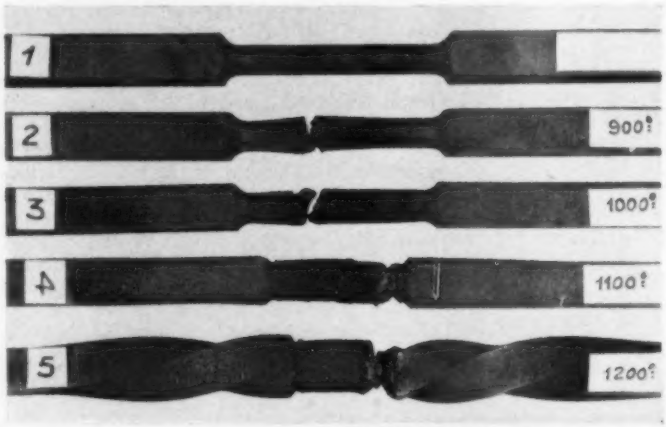
But it has been found that certain types of steel, particularly stainless steels with 17% Cr, in tests with free length, give rise not to a shortening but to an elongation of the test-piece (fig. 6). In this case the test with fixed length leads to an axial compression stress being opposed to the shearing stresses. The number of revolutions before fracture can then be notably higher in certain cases than with a free length, a phenomenon which is observed much less markedly with 18/8 steels.

It thus appears that the torsion test, according to the intrinsic tendency of the metal to elongate or contract in a greater or less degree when it is free to do so, may lead, in a fixed length test, to variable and complex stress conditions, which may possibly upset the expected results.

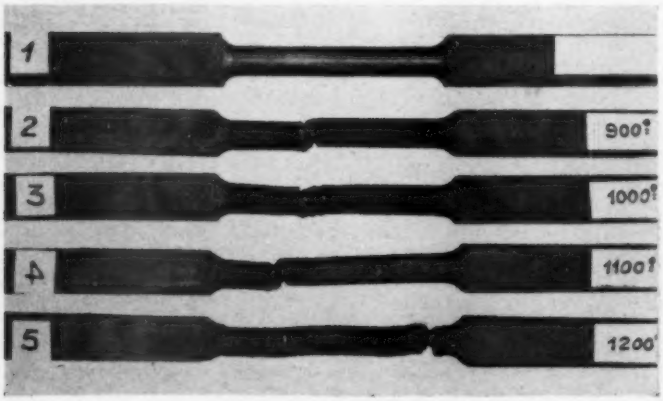
The obvious attraction of this method of testing largely lies in its simple quantitative character, the ductility of the metal being measured by the number of revolutions before fracture, this varying considerably according to the experimental conditions. This advantage certainly explains the interest which many steel formers have shown in it. In spite of this, the authors have not pursued the use of this experimental method very far, owing to the difficulties in principle mentioned above.

Tensile tests

Hot tensile tests have been less used than torsion tests in studying the forgeability of steel[1, 7, 8, 9]. The experimental arrangement used for hot tensile tests is shown in fig. 7. A testing machine using a



5 LEFT *Hot-torsion tests on* 18/8 *austenitic steel (Steel A). Note shortening and swelling of test-piece with rising temperatures.* 4 *and* 5 *have broken at set-downs*

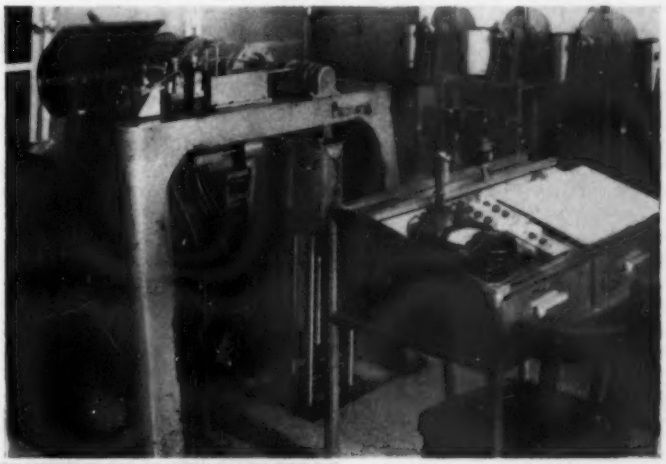


6 RIGHT *Hot-torsion tests on ferritic stainless steel* 17% *Cr (Steel B). Note elongation with rising temperature*

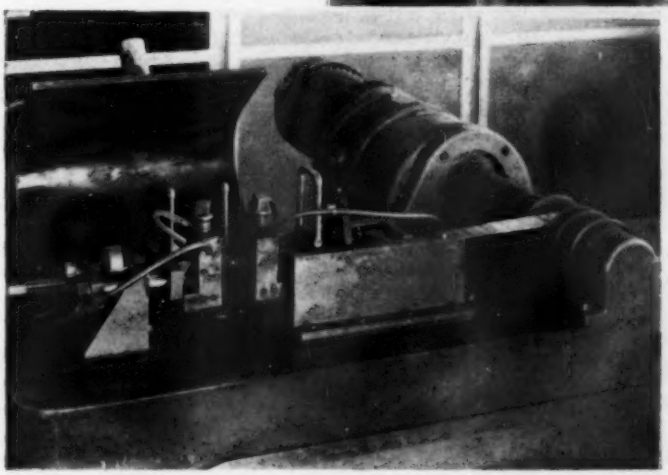
similar principle had already been put forward by R. Canard[8] in 1946.

The test-piece is made with a calibrated portion 8 mm. dia. × 5 mm. long ($\frac{5}{16}$ in. dia. × $\frac{3}{16}$ in. long), machined in the centre of a forged bar, 12 mm. square × 140 mm. long ($\frac{1}{2}$ in. square × $5\frac{1}{2}$ in. long). The test-piece is fixed in the machine heads by simple cotter-pins through two holes in each end of the square bar. Heating of the test-piece is obtained by Joule effect by means of a transformer supplying a current of high intensity and low voltage, which heats up the calibrated portion very quickly. In every case, intensity is regulated so as to ensure a rise to the required temperature in a constant time of 60 plus or minus 5 sec. Current is brought to the test-piece by two heavy pieces of copper, of which the movable top portion ensures contact by being screwed on. These two copper pieces are cooled by circulated water (fig. 8 (a)). The movable end of the test-piece is attached to a sliding carriage, of which the movement is controlled by a special high resistance metal strip wrapped round a drum (fig. 8). Deformation can be done at two widely differing speeds, one relatively moderate (about 500%/sec.), about the deformation rate of a rolling mill, and the other much higher (about 40,000%/sec.), which can be compared to that of a drop hammer.

In order to work at moderate speed, a dog clutch is used, at the moment the test-piece reaches the required temperature, to connect the movement of the draw-drum to the rotary axis of a speed reducer

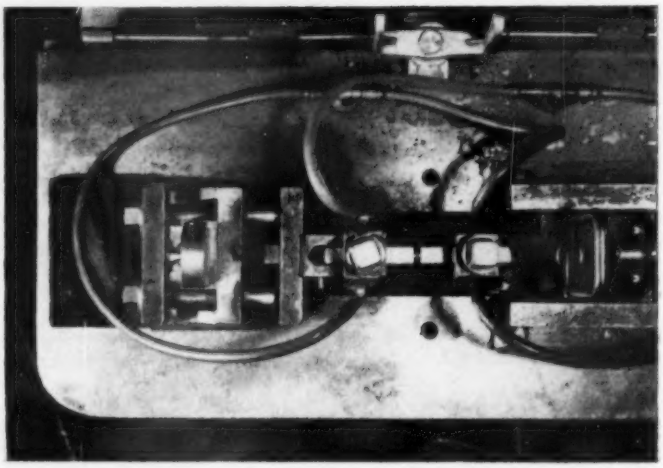


7 RIGHT *Hot-tensile test machine. In the foreground is control desk with optical pyrometer*

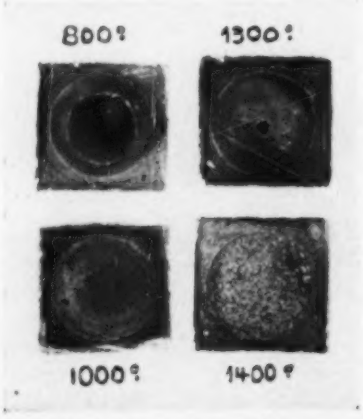


8 (a) LEFT *Hot-tensile testing machine*

8 (b) *Hot-tensile test machine as seen from above*



driven by an asynchronous motor. For working at high speed, an 80 kg. weight is released in free drop at the right moment. This weight, when it has reached a speed of 3 m. (10 ft.)/sec., carries the draw carriage along with it, through a face-plate, a second metal strip and a drum. Control of the clutch or of the falling weight, synchronized with the break of the heating current, is carried out automatically by an interrupter worked from a control panel. The operator carries out this movement at the precise moment that the test-piece comes up to the desired temperature, reading this by means of an optical pyrometer. A preliminary check has enabled the optical temperature to be corrected to the true temperature of the steel. Experience shows indeed that, even when using thin thermocouple wire and well-made welds, the flow of heat through the thermocouple wires is likely to cool down the thermocouple weld and so lead to errors. Welding of the thermocouple wires to the test-piece is, moreover, a difficulty which slows down any rapid development of the apparatus, particularly if it is to be used for control purposes in industry. Fig. 9 shows the appearance of fractures observed on test-pieces broken at different temperatures.



9 *End view of broken hot-tensile test-pieces*

This test has the following advantages: (*a*) Simple and very strong apparatus; (*b*) simple definition of stresses applied; (*c*) ease of adaptation for use with Joule effect heating, which allows of a very rapid temperature rise and of avoiding excessive oxidation of the steel; (*d*) wide range of test temperatures (from 800°C. to melting point of metal), see fig. 9; (*e*) great speed and simplicity in making test-pieces; (*f*) great speed of test (3 min. in all per test-piece); (*g*) use of two very different speeds of test corresponding to the deformation speeds of the chief types of working tools; and (*h*) simple quantitative indication of the ductility of the steel by determining the contraction or striction of test-piece.

It should be remarked, on this subject, that experience with this test has shown that measurements of diameter after breaking were easier to carry out and more precise than those of elongation of fracture. Moreover, it has been found that the contraction $\triangle\phi$, or the percentage reduction in diameter of the test-piece after fracture, proved in practice to be a more accurate indication of ductility in some cases than reduction in area, and this figure has been adopted for the whole of the investigations.

An objection to this test should be mentioned,

continued on page 263

High-conductivity copper castings

Production of copper castings with the highest attainable electrical conductivity has been achieved by suitable smelting methods. This translation of an article by E. Brunhuber in 'Radex Rundschau,' 1958, (7), discusses two-stage melting for dehydrogenation and deoxidation, and the relation between electrical conductivity and admixture concentration

THE PRODUCTION OF COPPER CASTINGS with a high degree of electrical conductivity is both a smelter's and a foundryman's problem. High electrical conductivity can only be attained with a high-purity melt. Technologically, this implies the use of chemically practically pure copper in the charge and suitable precautions against the introduction of any metallic contamination during the melting and casting processes.

In practice, this is not such an easy problem. Electrolytic copper satisfies high-purity requirements, as far as the charge is concerned, but always contains much hydrogen, which is the cause of excessive porosity in the resulting casting. The production of sound, electrically highly-conducting castings is, therefore, dependent on the adoption of a metallurgically-suitable melting routine, with the object of dehydrogenating a high-purity melt without introducing metallic contaminations.

Removal of hydrogen

Such dehydrogenation is most suitably effected by oxidizing the melt when the occluded hydrogen becomes replaced by oxygen. As a rule, it is sufficient for this purpose merely to run the melt with an oxidizing flame; the use of oxidizing fluxes would be equally correct but should not be necessary in the case of pure copper melts, providing an oxidizing flame can be maintained. Besides this, such fluxes may themselves contribute to contamination of the metal bath, if they should contain or form substances capable of going into solution in the molten metal. Since all oxidizing additions essentially consist of a fluxing medium and a metallic oxide, the composition and nature of such smelting aids should be particularly carefully watched. Even quartz sand, ordinarily used to increase the density of the slag blanket, may cause undesirable silicon enrichment of the casting metal with consequent loss of electrical conductivity.

In practice, therefore, not only are oxidizing fluxes avoided but no deliberate attempt is made to maintain a slag blanket over the metal bath. This includes the use of the otherwise very usual charcoal, which is often supplied in an insufficiently carbonized state, regarded from the standpoint of the smelter's requirements. As a rule, charcoal liberates hydrocarbon fractions; and also steam, particularly since it often exhibits traces of humidity.

The method of no-flux smelting in an oxidizing flame, dispensing with any kind of slag cover whatsoever, has, therefore, proved to be the most suitable. This, however, necessitates the use of reverberatory furnaces and the exclusion of sulphur-containing fuels such as coke or fuel oils, and particularly indicates the use of town gas or producer gas, as well as of desulphurized petroleum fuel oils. A very low sulphur content may in fact be harmless since the oxidizing melt also removed the sulphur by the formation of cuprous oxide by the equation:

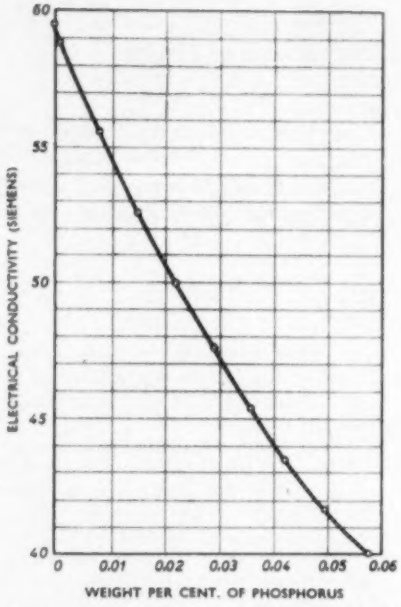
$$2Cu_2O + Cu_2S \rightleftharpoons 6\,Cu + SO_2$$

The increasing oxygen concentration, as already explained, displaces the hydrogen, but the oxygen remains in solution and must be removed. This so-called deoxidation is effected by the addition of an element having a higher affinity for oxygen than copper; the cuprous oxide Cu_2O formed in the melt is reduced and the reaction product is the oxide of the deoxidation agent used, which must therefore be easily capable of slagging.

In practice, phosphorus is usually employed for this purpose in the form of a 10—15% phosphor-copper alloy. To obtain a really highly-conductive copper casting, however, phosphorus must be avoided, since it has a deleterious effect on electrical conductivity.

Deoxidation

Complete deoxidation such as required to develop the optimum characteristics of the material in any case requires the addition of the deoxidizing agent in excess, so that the excess quantity now appears in solution in the copper as a fresh alloying constituent, and in this manner reduces the electrical conductivity. Since the hydrogen content of the



1 *Influence of phosphorus on the electrical conductivity of copper (after Pawlek and Reichel)*

cathode copper constituting the charge will always vary, there is no guarantee even when conducting the oxidizing melt under rigid control that any particular degree of oxidation can be repeatably maintained, hence there will be no reliable deoxidation unless an excess of the deoxidizing agent is used.

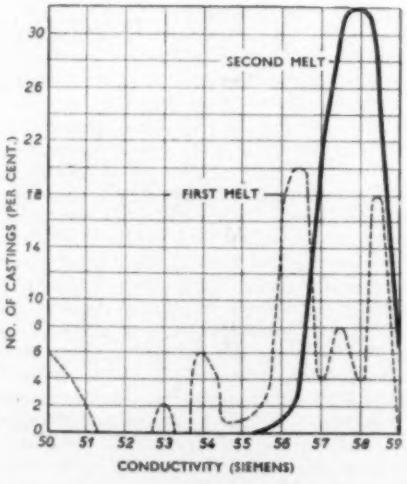
Hitherto it has been endeavoured to circumvent this difficulty by first thoroughly oxidizing the melt not only by running an oxidizing flame but also by an additional air blast. After this, ' poling ' is resorted to in the classical way until a residual oxygen content of 0·02 — 0·05% is attained, which can be controlled by regular sampling of the melt, from the appearance of the sample fracture (percentage of contained cuprous oxide). Then removal of the small remaining oxygen percentage can be performed with only a slight excess of deoxidizing agents. This enables high conductivity values of over 50 Siemens to be obtained, but only at the expense of fairly wide and irregular scattering of the values; the industrially guaranteed acceptance values are consequently restricted to 50—53 Siemens.

Oxidized copper which has not been poled but only once deoxidized furnishes similar but usually even more widely scattering values. It is more suitable, consequently, to use a process embodying two separate melting stages: a first melt run with a purely oxidizing flame, for expelling the hydrogen, followed by deoxidizing; and an oxidizing remelt for refining the contained deoxidizing agent.[2]

Double melting technique

It has been found in this regard that maximum conductivity values can be far more reliably obtained if admixtures to the melt are removed by oxidizing than if, conversely, the quantity of the deoxidizing agents is adjusted to obtain practically 100% oxygen elimination. The former process enables the increasing purity of the melt during oxidizing refining to be easily controlled by conductivity measurements on successive samples, while the opposite technique only enables purely visual control of the residual Cu_2O content with fairly wide scattering of the results.

In this way an already melted and deoxidized copper is used as the charge for the second melting; this produce is, however, more or less equivalent to so-called wirebar copper, so that it is found useful in making high-conductivity copper castings to start directly with deoxidized and phosphorus-free wirebars. Here again an oxidizing melt is run and serial conductivity tests used to control the progressive refining of the metal. After a conductivity value of 58—58·5 Siemens has been attained, the metal is cast without deoxidizing treatment. If the wirebars used for the charge contain oxygen, *i.e.* have not been deoxidized, the charge must be deoxidized as soon as it begins to flow freely and an oxidizing melt then run, as described earlier.



2 *Range of scattering of electrical-conductivity values measured in 36 copper castings, 1st and 2nd melts*

Deoxidants and resistivity

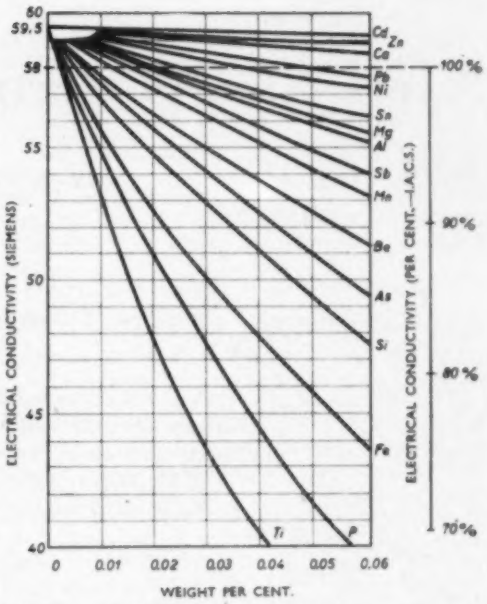
There remains therefore only the question of the selection of suitable deoxidizing agents, if non-deoxidized wirebars are to be used, or the first melt run with electrolytic copper as the starting material. The substances chiefly suitable for this purpose are the metals zinc, cadmium, calcium (in the form of calcium boride), magnesium, lithium and beryllium. Frequently, several of these metals are used together: the so-called duplex methods with magnesium and zinc, zinc and beryllium, calcium boride and lithium; or even triplex methods, *e.g.* zinc-beryllium-lithium, or other combinations.

Finally, general reference should be made to the conductivity behaviour of copper in the presence of admixtures. The maximum possible conductivity value for chemically pure copper (99·999% Cu) is, according to latest researches,[1] 59·5±0·15 Siemens, referred to a room temperature of 20°C., corresponding to a specific resistance of 1·681 ±0·004 ohm-cm. This has enabled a very important law to be discovered, *viz.* the linear relationship between specific resistance and concentration of admixtures. It applies, however, only on the condition that the particular alloying constituent is completely soluble, *i.e.* capable of forming mixed crystals with the copper. If this condition is not fulfilled, heterogeneous mixtures are produced and the straight-line resistance curve begins to deviate unsteadily.

Since in practical working, furthermore, there are as a rule not one but several such admixed constituents present simultaneously, the increases in resistance due to each of these in the range of concentrations of homogeneous, mixed-crystal formation are additive in effect. In engineering practice, specific resistance is often replaced by the reciprocal electrical conductivity values as a criterion. The hundredfold reciprocal value of the specific resistance is the Siemens unit of electrical conductivity. The conductivity curves of fig. 3, plotted against the admixture concentration, are thus in the range of homogeneous mixed crystal formation hyperbolic. Fig. 3 reproduces the results obtained by F. Pawlek and K. Reichel[1] supplemented by curves for the elements calcium, magnesium and titanium, obtained by the present author.

English-speaking countries use a different scale of representation, the so-called IACS (International Annealed Copper Standard) scale, the conversion factor being 100% IACS = 58 Siemens—at the time considered the highest attainable conductivity value for copper. On the basis of the research findings of Pawlek and Reichel, however, the optimum conductivity appears to be 59·5 Siemens = 102·6% IACS.

The so-called 'search coil' instruments[2, 4] are excellently suited for electrical conductivity measurements providing non-destructive, direct readings based on magnetic induction from any part of the surface of a casting.



3 *Influence of the more important alloying constituents on electrical conductivity of copper at 20°C.*

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(1) F. Pawlek and K. Reichel, Z. *Metallk.*, 1956, **47**, 347-356.
(2) F. Pawlek, K. Viessmann and H. Wendt, *Ibid.*, 1956, **47**, 357-363.
(3) F. Foerster and H. Breitfeld, *Ibid.*, 1952, **43**, 172-180.
(4) G. Bunge, *Ibid.*, 1954, **45**, 204-206.

Hot working of alloy steels

concluded from page 260

namely, that it does not lend itself to the study of practical processes of working, deformation in several passes being practically impossible with the type of machine adopted. Finally, it will be noted that this apparatus allows of the maximum resistance to deformation being determined, by a simple fitment, for each test-piece during each hot tensile test. To make this reading, the fixed end of the test-piece is connected, by means of two sliding rods, to a cross-member carrying a hard steel cone (figs. 8 (a) and 8 (b)). This comes into contact with a mild steel annealed disc and, during the tensile testing, makes an indentation of which the diameter depends upon the maximum stress imposed on the steel.

Application of electron microscopy

Precipitation of carbides and nitrides in a 12%-Cr low-carbon steel during tempering

JAROSLAV KOUTSKY and JAROSLAV JEZEK

Experiments to provide data on the course of carbide reactions in a low-carbon 12%-Cr steel during the breakdown of martensite and data on the breakdown of δ-ferrite during tempering are described. This article, translated from 'Hutnické Listy,' 1958, **13,** *(12), is the fifth of a series illustrating the role of the electron microscope in various metallurgical investigations*

IN THE THERMALLY HARDENED STATE the microstructure of steel AK 1 (CSN 17,021: C 0·15% max., Cr 11·5-14·5%) consists of martensite, or martensite and δ-ferrite, if the carbon content is very low and the chromium content is close to the upper limit.[1–4]

By tempering these steels at 315-425°C. the martensite is more readily grain refined as a result of the already marked manifestation of carbide precipitation.[5] As the temperature is increased, the tempered martensite visibly undergoes transformation into a sorbitic structure, the carbide particles become coarser, and the distances between them increase. If δ-ferrite is present in the tempered steel, when observed under an optical microscope at up to 650°C., the ferritic grain remains unchanged save for the precipitation of carbide particles along the grain boundaries. By tempering at 750°C. the precipitate makes its appearance over the whole of the δ-grain, with the exception of narrow fringes surrounding the grain boundaries.[5, 6] At even higher temperatures (close to the Ac_1 point) the precipitate within the ferritic grains is coarser, and its quantity is reduced. The precipitate within the δ-ferrite grains is for the most part likewise considered to be a carbide phase.[4, 6, 7] Only Rickett[5] asserts that this fact has not so far been clearly proved.

In the range of quenching temperatures, according to the Fe-Cr-C equilibrium diagram (Cr, Fe)$_{23}C_6$ is a stable carbide in the steels we investigated. The formation of this special carbide is dependent on the diffusion energy of the alloying element under the thermochemical conditions.

According to Kurdyumov[8] the breakdown of martensite during the tempering of carbon steels may be divided into two stages; in the first, which takes place gradually at normal temperatures, but even quite rapidly at 100-150°C., and which is designated as a di-phase or heterogeneous breakdown, there is formed a structure of tempered martensite composed of the alpha solid solution with the initial concentration of carbon and on the other hand of fine cementite particles, in the vicinity of which the solid solution is impoverished in carbon. Starting from temperatures of 150°C. and above the second stage in the breakdown of martensite occurs, in which the dominant part is played by carbon diffusion. The carbide particles grow, while the concentration of carbon in the solid solution is reduced from this point onwards; simultaneously coagulation of the carbide particles takes place.

The alloying elements have an influence on the carbon diffusion, and therefore affect the level of the temperature interval of the second stage in the breakdown of martensite. Additions of chromium, cobalt, tungsten and molybdenum raise this temperature level to 350-450°C. In the first stage of the breakdown of martensite in alloy steels, in view of the lack of diffusion energy of the alloying elements at these lower temperatures, cementite is

precipitated with a content of the alloying element corresponding to its concentration in the solid solution. With the development of the second stage in the breakdown of the martensite suitable conditions arise for the formation of special carbides.

Kuo[9] carried out an X-ray analysis of carbides which were electro-chemically isolated from heat-treated specimens of (0·30% C, 12·2% Cr) steel—heated at 1,100°C. for 30 min. and quenched in water and tempered at selected temperatures for 1 h. and cooled in air. He established the presence of these carbides: up to 500°C., Fe_3C; 525-625°C., Cr_7C_3; 650-670°C., $Cr_7C_3 + Cr_{23}C_6$; above 700°C. $Cr_{23}C_6$. These results obtained by Kuo, and likewise the results of a quantitative analysis of a substance isolated by the author[4] in AK 1 steel, and the course of the curves of static mechanical properties (changes in hardness and tensile strength) in relation to the tempering temperature confirm the validity of the facts stated above in relation also to 12% chromium steel with a low carbon content.

During an investigation of the isothermal breakdown of martensite in steel with a high carbon content in the zones in which the existence of the carbide $(Cr, Fe)_{23}C_6$ is both thermochemically and kinetically possible, Lashko and Nesterov[10] discovered the so-called gamma phase in the first stages of the breakdown; this is a metastable phase with an austenitic lattice and a high carbon content. Kuo on the other hand maintains that over the whole range of tempering temperatures, in which diffusion of chromium may occur, in high-chromium steels only one course is possible for the carbide reactions on the lines: $Fe_3C \rightarrow (Cr, Fe)_7C_3 \rightarrow (Cr, Fe)_{23}C_6$. The γ-phase is considered to be residual austentite, which was not dissolved during electrolytic isolation.

The mechanism of the formation of special chromium carbides is presented by Balluffi and others[11] and by Lashko and Nesterov as follows: cementite having reached the equilibrium concentration with the surrounding solid solution, is precipitated in the latter, and the carbide $(Cr, Fe)_7C_3$ is formed from free nuclei in the solid solution. In view of the fact that cementite can be enriched with chromium up to about 18%, it is not likely that $(Cr, Fe)_7C_3$ would be formed directly from the impoverished solid solution. Kuo assumed that the $Fe_3C \rightarrow Cr_7C_3$ transformation takes place directly from the reaction between Fe_3C and the surrounding solid solution. Wever and Koch,[12] by a detailed study of an isolate from steel with a chemical composition of 0·4% C and 3·5% Cr by means of an electron microscope, showed that during the isothermal breakdown of austenite the carbide $(Cr, Fe)_7C_3$ is formed directly from the cementite lamellae.

This work therefore confirms the opinion of Kuo. He investigated[13] by means of an electron microscope the formation of Cr_7C_3 in medium-alloy chromium steels which were tempered at various temperatures. He discovered that the carbide $(Cr, Fe)_7C_3$ has a dendritic structure which is fundamentally different from the large cementite formations, but the conclusion which follows from the photographs which he presents concerning the direct $Fe_3C \rightarrow (Cr, Fe)_7C_3$ transformation is not as unambiguous, as that in the work of Wever and Koch.

Heiskanen[14] used an electron microscope to investigate carbide isolates of steel containing 0·18% C and 12·7% Cr, which were formed during long-term tempering at 500-550°C. after quenching in oil from 1,025°C. (holding time 2 h.). The electron micro-photographs provide objective documentation of the growth in precipitation as a function of time and temperature, but the quantitative differences in the carbide phases are not too clearly apparent on them. The results of the X-ray analysis confirm the sequence of the carbide reactions which take place during the formation of the stable carbide in tempered high-carbon steel, as propounded by Kuo. During the course of tempering secondary hardening is not revealed, which is evidence of the direct formation of the chromium carbide from cementite.

We have carried out the following experimental programme with the aim of providing authentic data on the course of the carbide reactions in AK 1 steels, which occur during the breakdown of martensite, and also of obtaining more detailed information concerning the change in, and breakdown of, δ-ferrite during tempering.

Experimental materials and methods

As material for the experiments we made use on the one hand of specimens of mechanically worked steels from current production melts of the following percentage compositions:

	C	Mn	Si	P	S	Cr	Ni	N
Melt A	0·15	0·36	0·21	0·022	0·013	13·40	0·14	0·026
Melt B	0·13	0·27	0·18	0·021	0·013	13·20	0·19	0·024

and on the other hand of specimens from a 5-cm. thick plate, which was sand cast; its chemical composition (%) was: 0·09 C, 0·50 Mn, 0·64 Si, 0·017 P, 0·030 S, 13·30 Cr, 0·40 Ni and 0·023 N.

After thermal hardening at 1,000°C. for 2 h. and quenching in oil, the structure of steel A is formed exclusively of martensite, that of steel B of martensite with a small amount of δ-ferrite, and that of steel C of martensite with up to 20% δ-ferrite.

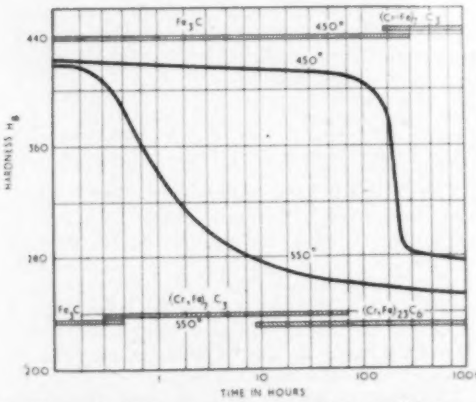
Specimens of all the melts were tempered by holding at a temperature of 750°C. for 2 h. and cooling in air; in addition specimens from melt A

were heat-treated for long periods (up to 1,000 h.) at temperatures of 450 and 550°C. Changes in the microstructure of the specimens which were tempered by short heating periods were examined with the optical and the electron microscope. For the study of the microstructure with the electron microscope extraction replicas were drawn from the specimens. In some instances the very fine particles which were captured in the extraction replica were subjected to electronographic analysis; if the precipitate was coarse, the particles in the extraction replica were subjected to X-ray analysis.[15] On the specimens which were subjected to long-term heat treatment studies were also made of the changes in hardness as a function of the time of heat treatment, and of the qualitative composition of the isolate. The precipitate was separated from the solid solution by chemical isolation in a 10% solution of bromine in methylalcohol. (This method was tried out by Cihal[21] during the isolation of carbides in 18/8 austenitic steel.)

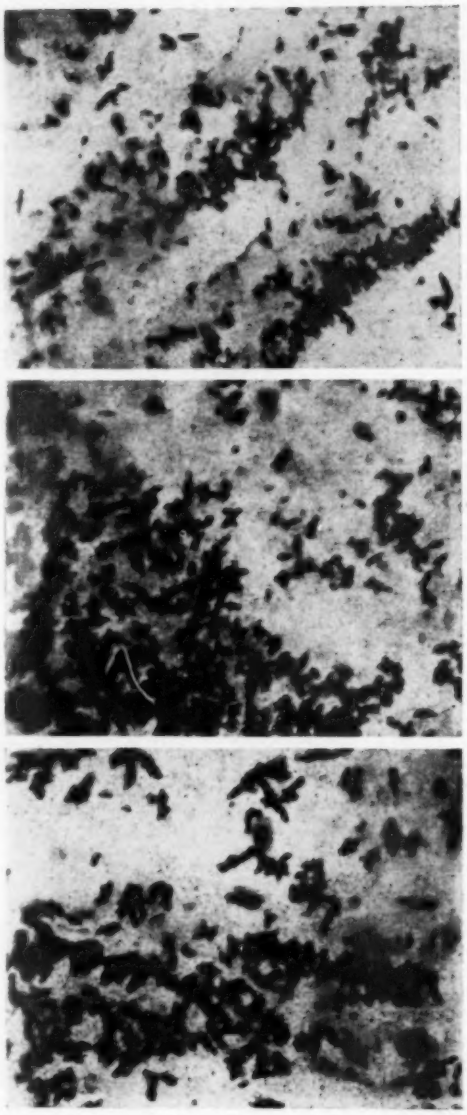
Results of the experiments

The data which were obtained from the study of the specimens with the optical microscope, agree with the results which have already been published elsewhere concerning these steels, and which we have described in the introduction; we shall therefore make no mention of them in greater detail.

In specimens of steel A, after tempering at 450°C. for up to 200 h., the carbide phase consists of cementite (fig. 1). After a period of 200-300 h. the lines of the carbide $(Cr, Fe)_7C_3$ appear on the X-ray diagram. Within these limits of the heating period there is also a marked drop in hardness. After heat treatment for 1,000 h., by the X-ray diagram



1 *Changes in hardness and structure of the carbide phase during long-term tempering of Steel A at temperatures of 450—550°C.*

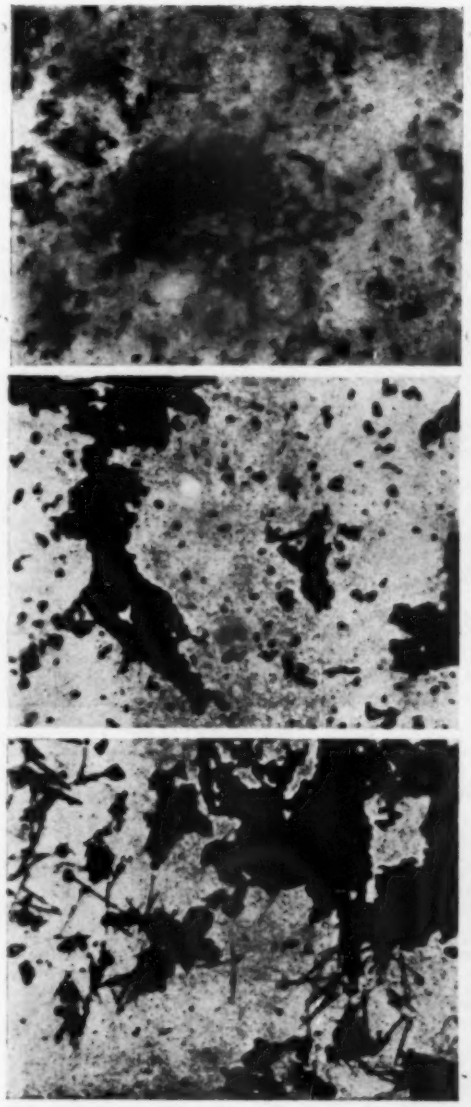


Extraction replicas × 10,000

2 TOP *Steel A, 1,000°C./oil—450°C./30 min./air*

3 MIDDLE *Steel A, 1,000°C./oil to 450°C./100 h./air*

4 BOTTOM *Steel A, 1,000°C./oil—450°C./100 h./air*

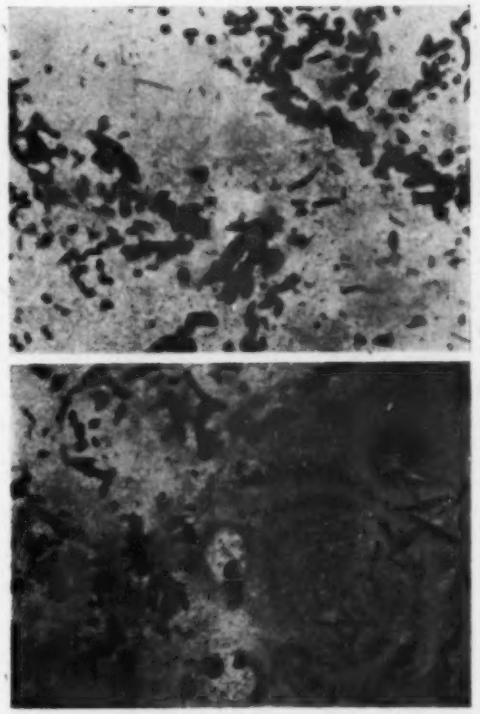


Isolates × 10,000

5 TOP *Steel A*, 1,000°C./*oil*—550°C./30 *min.*/*air*

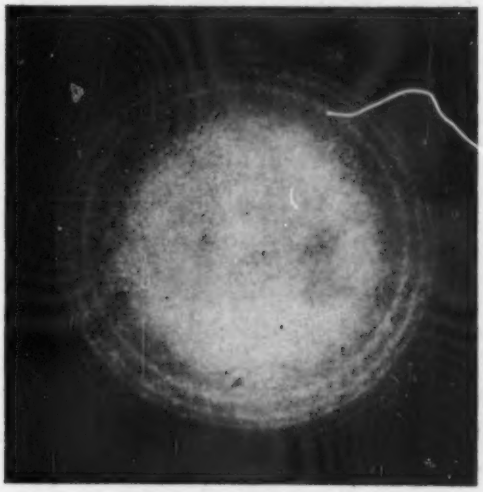
6 *Steel A*, 1,000°C./*oil to* 550°C./1,000 *h.*/*air*

7 *Steel A*, 1,000°C./*oil*—550°C./1,000 *h.*/*air*

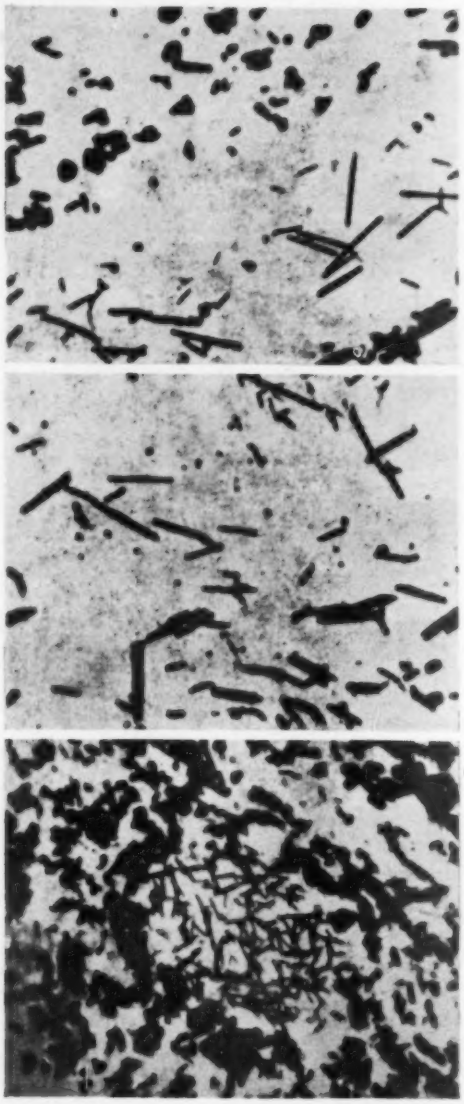


Extraction replicas × 10,000

8 TOP *Steel A*, 1,000°C./*oil*—550°C./16 *h.*/*air*
9 *Steel A*, 1,000°C./*oil*—550°C./1,000 *h.*/*air*



10 *X-ray diffraction of fig.* 8

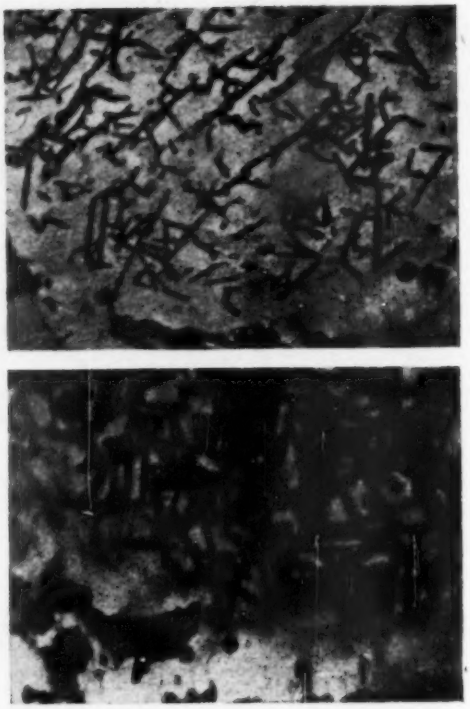


Extraction replicas × 10,000

11 TOP *Steel A*, 1,000°C./*oil*—650°C./20 *h./air*

12 *Steel A*, 1,000°C./*oil*—750°C./2 *h./air*

13 *Steel A*, 1,000°C./*oil*—650°C./2 *h./air*



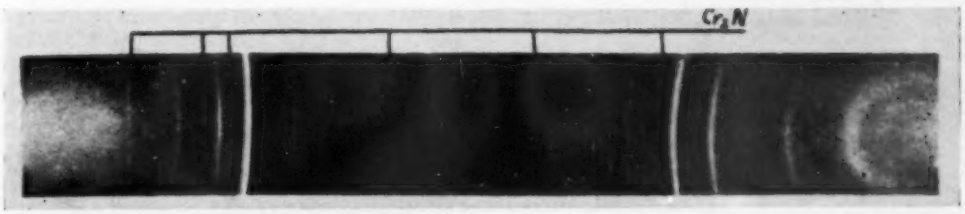
Extraction replicas × 10,000

14 TOP *Steel B*, 1,000°C./*oil*—650°C./20 *h./air*

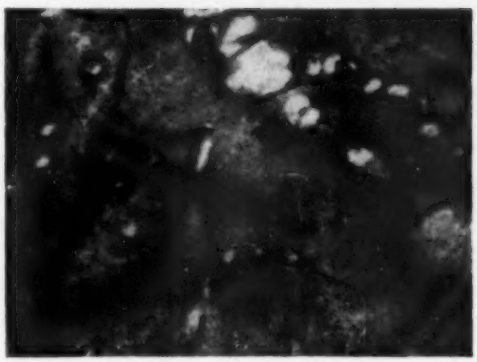
15 *Steel B*, 1,000°C./*oil*—750°C./2 *h./air*

Picric acid etch × 300

16 *Steel C*, 1,000°C./*oil*—750°C./2 *h./air*



18 *X-ray diagram of fig.* 17



Extraction replica × 10,000

17 *Steel C, 1,000°C./oil—750°C./2 h./air*

we can detect the presence of the carbide $(Cr, Fe)_7C_3$, and also of what is possibly a small quantity of the carbide $(Cr, Fe)_{23}C_6$. After the shorter and the longer periods of heat treatment, the carbide particles have for the most part the formation of imperfect, stellate particles or short dendrites with small side shoots, such as we see in figs. 2 to 4, which present photographs of extraction replicas of specimens after 30 min., 100 and 1,000 h. heat treatment at 450°C.

At a temperature of 550°C., in the initial stages of the breakdown of martensite, cementite is also present in the isolate, which as the time of heating is increased is transformed into $(Cr, Fe)_7C_3$; and at an even later stage this last becomes $(Cr, Fe)_{23}C_6$. After heat treatment for longer than 100 h. it is, moreover, possible to identify the lines of the carbide $(Cr, Fe)_{23}C_6$ on the X-ray diagram (fig. 1). The hardness, having fallen appreciably in the first stage of the heat treatment, in contrast only falls quite gradually after longer periods.

On this occasion we also examined the morphologies of precipitates in a chemically isolated fraction which was dispersed in benzene and applied to a collodion blank. From figs. 5 and 6 we can establish only that there is coarsening of the precipitate, as the duration of the heat treatment is increased; we cannot distinguish any qualitative differences in the appearance of the precipitate. After 1,000 h. heat treatment, in one instance we established in an isolated deposit the presence of acicular particles which were markedly different from the basic carbide precipitate (fig. 7). On the X-ray diagram of the isolate of this specimen, apart from the lines of the carbide $(Cr, Fe)_{23}C_6$, there are also visible very strong lines of the nitride Cr_2N (Table I).

TABLE I *Results of the X-ray analysis of the isolate shown in figs.* 6 *and* 7

Measured distances Å	Intensity	Tabulated distances $Cr_{23}C_6$ Å	Tabulated distances Cr_2N Å
3·32	Very weak		
3·19	Very weak		
2·65	Very weak		
2·54	Very very weak		
2·38	Strong	2·38	2·38
2·17	Weak	2·17	
2·11	Weak		2·10
2·07	Very weak		
2·04	Very strong	2·044	
1·83	Strong	1·88	
1·80	Strong	1·80	
1·78	Medium	1·77	
1·69	Weak	1·68	
1·62	Weak	1·62	1·63
1·61	Medium	1·60	
1·49	Very weak	1·49	
1·39	Very weak		1·38
1·33	Medium	1·33	
1·29	Medium	1·29	
1·26	Strong	1·25	1·26
1·23	Strong	1·23	
1·19	Medium strong	1·19	
1·18	Medium	1·17	

From a study of the extraction replicas of these specimens it follows that this acicular precipitate makes its appearance in a very fine form and in an appreciable quantity, even after short periods of heat treatment (fig. 8). From this extraction replica a diffraction diagram (fig. 10) was obtained, on which were identified the lines shown in Table II. Apart from the diffraction lines of the carbide $(Cr, Fe)_{23}C_6$, the lines of the nitride Cr_2N are also to be

TABLE II *Evaluation of the diffraction diagram of the extraction replica shown in fig. 8*

Measured distances A	Tabulated distances $Cr_{23}C_6$ A	Tabulated distances Cr_2N A
2·38	2·38	2·38
2·16	2·17	
2·10		2·10
2·04	2·044	
1·88	1·88	
1·80	1·80	
1·68	1·68	
1·62	1·62	1·63
1·60	1·60	
1·48	1·48	
1·40		
1·38		1·38
1·29	1·29	
1·25	1·25	1·26
1·21	1·23	

seen. As the duration of the heat treatment is increased, this acicular precipitate grows. As may be seen from figures 8 and 9, the acicular particles are precipitated predominately within the acicular crystals of residual martensite, whereas the carbides are precipitated predominately around their periphery. At higher tempering temperatures, just like the carbide particles, whose forms become simplified so far as their ordering is concerned, so likewise the acicular particles are to be found in certain locations in a coarse form (figs. 11 and 12).

In the specimens of steel B, in whose microstructure δ-ferrite is to be found, by means of the electron microscope it was possible to establish the precipitation of these acicular crystals within the ferritic grains after as little as 2 h. heat treatment at 650°C. (fig. 13). Within the grains of δ-ferrite very fine and long needles are formed, from the orientation of which it is apparent that they were precipitated along the crystallographic valleys of the ferrite. If we prolong the time of heat treatment or increase its temperature (figs. 14 and 15), the particles increase in size. Their structure and the transformations through which they pass as a result of various methods of heat treatment are analogous to the formation and changes of the acicular crystals of sorbite. In view of the small number of islands of δ-ferrite in the structure of this steel, it is impossible to identify the precipitate with reliability since it makes its appearance within the ferrite grains.

In fig. 16 is shown the microstructure of a specimen of steel C after hardening and tempering: it was held at 1,000°C. for 2 h. and quenched in oil and held at 750°C. for 2 h. and cooled in air. Apart from sorbite there is in the structure up to 20% δ-ferrite, in which a very coarse precipitate is already visible after this short period of heat treatment. From this specimen we took an extraction replica (in this instance it is shadowed with carbon), and in the δ-ferrite there became visible acicular particles which were already strongly developed and had the same morphology and size as the particles in the specimens of steel which had been mechanically worked, but was heat treated in the same manner (fig. 17). The greater quantities of δ-ferrite and the coarser precipitate inside the ferritic grains afforded a real opportunity to identify this precipitate by X-ray analysis. The X-ray diagram produced from this extraction replica (fig. 18) indisputably shows the presence of the nitride Cr_2N as well as the carbide $(Cr, Fe)_{23}C_6$ (Table III).

TABLE III *Evaluation of the X-ray diagram (fig. 18) of the extraction replica of the specimen shown in fig. 17*

Measured distances A	Tabulated distances $Cr_{23}C_6$ A	Tabulated distances Cr_2N A
2·38	2·38	2·38
2·23		2·22
2·17	2·17	
2·12		2·10
2·04	2·044	
1·88	1·88	
1·80	1·80	
1·77	1·77	
1·68	1·68	
1·62	1·62	
1·60	1·60	
1·49	1·49	
1·48	1·48	
1·39		1·38
1·33	1·33	
1·29	1·29	
1·265		1·26
1·25	1·25	
1·23	1·23	
1·19	1·19	
1·17	1·17	

Discussion of the results

The results of our experiments confirm the course of the carbide reactions during the breakdown of martensite in 12% chromium steel as outlined by Kuo: $Fe_3C \rightarrow (Cr, Fe)_7C_3 \rightarrow (Cr, Fe)_{23}C_6$. The absence of anomalies on the hardness curves is evidence of the fact that the most stable of these carbides is not formed from independent nuclei by precipitation from the solid solution, but by the reaction of the less stable carbide with the surrounding solid solution. The carbides are precipitated predominately around the periphery of the original martensitic acicular crystals. Their extremely well-ordered initial form, which becomes simpler with the passage of time and with a high heat-treatment temperature, likewise indicates indirectly the mechanism of direct transformations during the carbide reactions. In a similar way to Heiskanen,[14] during these studies we also did not

establish whether the various crystallographic structures of the individual types of carbides likewise correspond to a specific internal structure, and moreover a study of the morphology of the carbides alone cannot give any idea, which form of the carbides is present at a particular stage of the breakdown of martensite.

The acicular phase is precipitated in the sorbite within the acicular crystals of residual martensite along the cleavage planes of the ferrite. As the temperature and duration of tempering increase, the acicular crystals grow. The precipitate which makes its appearance within the δ-ferrite grains has the same nature. With the growth in the temperature and duration of tempering, it passes through the same transformations as the acicular precipitate in the sorbite. On the boundaries of the ferritic grains, where the carbides are precipitated, this precipitate does not appear. Electron diffraction of an extraction replica taken from a specimen with a sorbitic structure containing a large quantity of fine acicular crystals, and X-ray analysis of an extraction replica of the structure with a very large quantity of δ-ferrite and markedly developed acicular crystals, showed that the acicular phase, which makes its appearance in sorbite and δ-ferrite, has the same crystallographic structure, and that this phase is the nitride Cr_2N. (In this nitride there may be partial substitution of the chromium atoms by an iron atom, and its formula may be written as $(Cr, Fe)_2N$.) The morphology of this precipitate and the way in which it is precipitated is, therefore, governed by its crystallographic structure.

In order that we might determine the thermochemical conditions governing the precipitation of nitrides in 12-13% Cr steel in relation to their content of nitrogen, we carried out the following analysis. The free enthalpy of the reaction of the diffusion of nitrogen in alpha iron in accordance with the equation: $Fe_{(\alpha)} + N_2 = 2\,(N)$, on the assumption that 0·01% nitrogen was diffused into the alpha iron is given by the equation.[17]

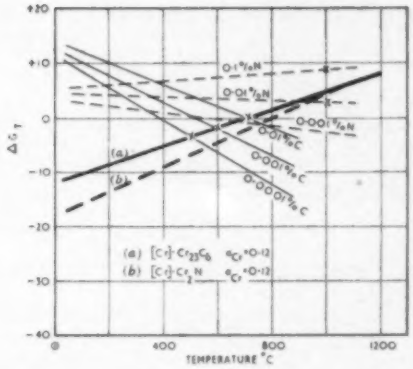
$$\triangle G_T = 9{,}760 - 2{\cdot}3T \text{ cal},$$

where T is the absolute temperature in °K.

This equation we may establish for various concentrations of nitrogen, on the assumption that we know the equation for the partial free enthalpy, and so construct the graph of the rectilinear relationship between the chemical potential of the nitrogen in the solid solution of alfa iron and temperature at various concentrations of nitrogen.

The formation of chromium nitride (Cr_2N), which is stable at high temperatures, takes place in accordance with the equation: $4\ Cr + N_2 = 2\ (Cr_2N)$ (this nitride is not stechiometric). The free enthalpy of this reaction is then similarly:

$$\triangle G_T = -44{,}000 + 24{\cdot}0\ T \text{ cal}.$$



19 *Changes in chemical potential of carbon and nitrogen in alpha iron, and of chromium carbide and nitride in relation to temperature for* 12%*-Cr steel*

In accordance with this equation we can construct the graph of the rectilinear relationship between the free enthalpy of this nitride and temperature for a given concentration of chromium in the solid solution.

In a similar manner we express the change in the free enthalpy of the reaction of the formation of the stable chromium carbide:

$$\frac{23}{6}Cr + C = \frac{1}{6}Cr_{23}C_6 - \triangle G_T = -16{,}380 - 1{\cdot}54\,T$$

and the change in the chemical potential of the carbon diffused into the alpha iron at various concentrations of carbon:[18]

$$C = (C)\alpha$$

$0{\cdot}01\%\quad \triangle G_T = 20{,}000 - 25{\cdot}98\ T$
$0{\cdot}001\%\quad \triangle G_T = 20{,}000 - 30{\cdot}55\ T$ etc.

The calculated course of all these rectilinear functions plotted against G and T co-ordinates are given in fig. 19.

The concentrations of nitrogen and carbon in the martensite and δ-ferrite phases in steels B and C were calculated by means of the equations given in the work of Andrews. (More detailed descriptions of the equations used are set out in our most recent work.[20])

The total content of carbon in the steels in which we investigated the breakdown of δ-ferrite was 0·13% in the steel designated B, and on the other hand 0·09% in steel C. If we take into consideration that, after tempering, in steel B there was 10% δ-ferrite and in steel C 20%, we arrive by means of calculation at the carbon concentrations in the

TABLE IV *Carbon and nitrogen concentrations in two steels*

	C%		N%	
	20% *ferrite*	10% *ferrite*	20% *ferrite*	10% *ferrite*
Total	0·09	0·13	0·02	0·02
Austenite	0·109	0·142	0·023	0·0214
Ferrite	0·013	0·017	0·007	0·0077

martensite and in the δ-ferrite which are given in Table IV.

Let us examine the concentration of carbon in the δ-ferrite in greater detail. From fig. 19 it follows that ferrite with a 0·01% carbon content is theoretically in equilibrium with the chromium carbide at a temperature just a little above 700°C. As the concentration of carbon in the ferrite of our particular steels is in both instances above 0·01% (0·013 and 0·017%), it follows from the diagram (fig. 19) that at the carbon concentrations calculated by us precipitation of the chromium carbides from the δ-ferrite would be thermochemically possible up to a temperature of about 750°C.

The nitrogen content in both the steels did not fall below 0·02%. Whereas 10 and 20% δ-ferrite are respectively present in the structure of the tempered specimens, by calculation we arrive at concentrations of nitrogen in the ferrite of 0·007 and 0·0077% (Table IV). From fig. 19 it is evident that at this concentration of nitrogen in the ferrite the latter is saturated with nitrogen, and precipitation of nitrogen from the ferrite must take place at temperatures below about 900°C.

If we compare the chemical potentials of the carbon and nitrogen diffused into the alpha iron, with the chemical potentials of the carbide and nitride of iron at a chromium concentration in the solid solution equal to 12%, we shall see that at these calculated concentrations precipitation of the nitrides from the δ-ferrite has definite preference from the thermochemical aspect. If we further take into consideration the fact that in our diagram for lack of thermochemical data we could not take into account the influence of the diffused chromium on the activity of the carbon and the nitrogen in the ferrite, and that the presence of chromium will obviously reduce their activity, also in addition that δ-ferrite with the calculated concentration of carbon, at a temperature above 600°C. for instance, will be a more stable phase than the chromium carbide, these facts alone may serve to emphasize the thermochemical preference of the precipitation of chromium nitrides from the δ-ferrite. Kinetic factors cannot play a part here, for in both instances precipitation is controlled by the diffusion of chromium.

Our thermochemical considerations are loaded with further inaccuracies, which are confirmed in other works,[17, 18] and to which attention has already been drawn in our literature also;[20, 21] the conclusions drawn from our reflections are therefore only very approximately valid. Nevertheless they do show, however, good agreement with the facts observed.

Conclusions

The results of our experiments and their comparison with those already published enable us to offer the following conclusions about the processes which we investigated:

Precipitation of carbides (1) Our studies confirmed the course and the mechanism of the carbide reactions during the breakdown of martensite in steel with 12-13% Cr and a low C content; they were carried out and described for a steel close to the composition used by Kuo and Heiskanen. (2) The individual types of chromium carbides do not, during this precipitation process, have their own specific structures by which one could be distinguished from another.

Precipitation of nitrides (1) By the use of special physical and chemical methods we were able: (*a*) to show the orientation and distribution of the acicular precipitate in martensite (or sorbite) and δ-ferrite; (*b*) to discover the relationship between its precipitation and growth on the one hand and the time and duration of tempering; and (*c*) to determine the crystallographic structure of the precipitate. (2) The acicular precipitate which is diffused from the martensite (sorbite) and δ-ferrite is chromium nitride (Cr_2N). (3) This nitride of chromium is identical with the precipitate which is formed within the grains of δ-ferrite. Our experiments showed that the doubts raised by Rickett concerning the carbide basis of this phase, which is precipitated from δ-ferrite during the breakdown of the ferritic phase, were justified. (4) Thermochemical considerations show that for the precipitation of the nitrides at temperatures below the Ac_1 point comparatively small quantities of diffused nitrogen are sufficient, and likewise that it is impossible to rule out the existence of some relationship between the brittleness of AK 1 steel at working temperatures and the migration of nitrogen.[22]

Acknowledgments

The authors esteem it their pleasant task to thank J. Neid of ZVIL, Pilsen, for collaboration in the X-ray analyses, and Vl. Cihal of SVÚOM, Prague, for valuable comments on the thermochemical analysis.

***The references** for this article will be given next month with the sixth of the series on electron microscopy.*

Tantalum

The high resistance to corrosion of tantalum makes this metal important in many specialized industrial applications. The following brief survey of some of its characteristics and uses is the sixth of a series on the newer metals appearing in this journal

WITH A RESISTANCE TO CORROSION comparable to that of platinum, the metal tantalum, despite its retractable working properties, is now being used in both the chemical and electrical industries as well as proving an ideal metal for surgery. A white metal, in appearance rather like platinum but much cheaper, tantalum has been used for some time in jewellery and in rayon spinnerets in the United States. Yet although such applications have not been developed in Britain, other more prominent uses have increased the versatility of the metal. One important example is the place taken by tantalum as a ' biologically-acceptable ' metal, in the form of fine gauze, thin plates and wire for making sutures, repairing damaged tissues and skull injuries.

Tantalum, though a modern metal, has a long history. The bicentenary of the opening of the British Museum brings to mind how a mineral among the many bequeathed by Sir Hans Sloane began the story of both tantalum and its sister element columbium or niobium. A sample of columbite sent to Sloane by John Winthrop, grandson of a governor of Connecticut, lay in the museum for decades before Charles Hatchett, son of a prosperous Long Acre coach-builder and brilliant chemist, decided to analyse it. Hatchett discovered a new element, and duly christened it ' columbium ' since it was in an American ore; hence the insistence by America on ' columbium ' rather than ' niobium ' for the modern metal. Columbium and tantalum are so closely alike in properties that Hatchett could be excused for not finding tantalum present as well, this before this English chemist was lured from his chemistry ' by the baneful effects of wealth ' as historian Thomas Thomson put it. Both tantalum and columbium are found in ferrous columbate-tantalate, Fe $(Cb\text{-}TaO_3)2$, the mineral being called ' columbite ' when columbium is in excess and 'tantalite' when tantalum predominates.

Just 150 years ago Wollaston, father of powder metallurgy who made malleable platinum so successfully, failed to improve on Hatchett's discovery of 1801; for he analysed the ore and simply concluded that columbium and tantalum are identical or the sole ' new ' metal present. By 1846, however, Heinrich Rose made a full study of columbites and tantalites from both America and Europe, and extracted from them two acidic oxides which were not identical. He called one of them niobic acid, and named the element ' niobium.' In 1802 Ekeberg, a Swedish chemist, had analysed a specimen of tantalite and named ' tantalum ' as a new element; hence it only remained for some ingenious chemist to find a method of separating the two elements by a practical method.

Extraction processes

It was in 1866 that Marignac published a classic method of separation, one so efficient that it is still in use today. The separation depends on the relative insolubility of potassium fluotantalate in comparison with the soluble nature of potassium fluooxycolumbate. In modern extraction processes using alkali, the ore is fused with caustic soda or potash in preference to the alternative method in which fusion with potassium or sodium bisulphate followed by washing with hydrochloric acid is used. After fusing with the caustic soda (with some sodium carbonate) the mass is extracted with water, and then with hydrochloric acid to remove iron and manganese and to leave hydrated oxides of tantalum and columbium, these being now dissolved in hydrofluoric acid and a potassium salt added to form the insoluble potassium fluotantalate. By further recrystallizing from dilute hydrofluoric acid the tantalum-bearing precipitate in purified in readiness for extracting tantalum powder.

Not until 1903 was the first ductile sample of tantalum prepared by Werner von Bolten of the well-known Siemens-Halske firm in Berlin. This eminent German metallurgist was searching for lamp filaments in days when every conceivable metal was being tried in place of the carbon filament of Edison and Swan. More attractive than osmium metal for this purpose, especially since osmium oxide is highly toxic, was tantalum, so much so that some hundred million lamp filaments had been produced from it before tungsten dominated the field. Although tantalum filaments can only be

used with direct current since crystallizing of metal occurs with a.c., tantalum lamps still prove useful on railroads where excessive vibration may be encountered.

Von Bolten's process, used by Siemens and Halske today, consisted in reducing the potassium fluotantalate by metallic sodium in steel containers, alternate layers of the tantalum compound and of sodium being placed in the vessel. After covering the large crucible with a lid the temperature is raised to begin the exothermic action, the mass after completion of this reaction and cooling yielding tantalum particles in a fused mass which is dissolved with dilute acid.

While German extraction processes have been based on this sodium method, American producers like the Fansteel Corporation prefer the electrolytic process, this after C. W. Balke as pioneer of all tantalum research considered the sodium method unsuitable for industry. In this alternative a cast-iron pot with graphite anode is used, a fused bath of potassium fluotantalate being electrolysed while tantalum oxide is added to the melt. Crystalline aggregates or ' trees ' of tantalum metal are formed, the mass left at the end being crushed, ground, and the dispersed metal separated by tabling. A final washing with dilute acid removes the remaining salts.

Methods of working

To convert tantalum powder into the massive metal an efficient high-vacuum working is essential since the metal has a high affinity for oxygen and nitrogen. With a melting point of 2,996°C. and an affinity for carbon, the choice of refractories proved difficult but this problem has now been overcome, the metal being sintered in vacuum like tungsten or molybdenum before compressing into bars up to 2½ ft. in length and 1 in. cross section. Such bars are gripped between water-cooled terminals in the vaccum furnace and then heated by passing a current through the bars. In the Siemens-Halske method the vacuum furnace has a number of tungsten rods as resistors for heating the tantalum by radiation. This German processing is more difficult than with tungsten or molybdenum and is more prolonged than the Fansteel sintering by direct electrical heating.

After sintering the tantalum bars are rolled, forged and further heated before completing the conversion to metal free from porosity. Forging, rolling and swaging in the cold are less difficult than might be expected, heating being forbidden because of oxygen absorption. Annealing in a vacuum furnace at an appreciable temperature is, however, always necessary before further reduction into sheet, which is now obtainable down to 0·0002 inches in thickness. For drawing to rod or wire metal carbide dies are used or diamond dies for fine dimensions.

Tantalum articles or parts can be chemically cleaned by use of hot chromic acid solutions, but grit blasting must be by steel grit to enable particles of grit to be removed with concentrated hydrochloric acid. Carbon tetrachoride is quoted as the most suitable cutting-medium, a liberal flooding being maintained during operations. Resistance welding and arc welding techniques have been fully developed in the Fansteel research laboratories.

Applications in industry

Modern applications of tantalum illustrate the scope for fabricating into a wide variety of components. The metal is supreme in corrosion resistance in chemical industry, very thin layers being fully resistant with correspondingly high heat transfer. Tantalum chemical plant may be used with wet or dry chlorine, bromine and iodine, with wet or dry salts except those which hydrolyse to strong alkalies, with concentrated acids except hydrofluoric acid and oleum or phosphoric acid at high temperatures, and with organic compounds except those which contain fluorine or which may form caustic alkalies.

In coil and bayonet heaters in the chemical industry the metal is ideal, annealed tantalum sheet of 0·01 inch thick having a tensile strength of 50,000 lb./sq. in., while worked sheet has a strength of 110,000 lb./sq. in. High steam pressures may thus be used, with heaters evaporating 200-lb. liquid per hour from dilute acid for a heating surface of only 0·5 sq. ft. Absorption systems for hydrochloric acid incorporating thin-walled tantalum tubes and a falling film system can yield, for example, 36% concentrated acid with cooling water at 87°F. Much old HCl plant has been written off with tantalum plant displacing such equipment.

The metal has been widely adopted in nozzles for chlorinating water supplies, in tantalum extruders for sausage casings and other products, in fountain-pen nibs, and in the United States in spinnerets for rayon production.

In chemical plant in which tantalum is used it is particularly in heat transfer that the metal has advantages over other corrosion-resisting materials. In the nitric acid industry a heater with an effective heating area of 2·36 sq. ft. and using 150 lb. of steam, for example, can evaporate a ton of 95% acid per hour. Multiple bayonet heaters and heat exchangers with as many as 174 seamless tubes of tantalum bring a wide demand for tantalum tubing, which is available as lap-welded, inert gas arc welded and seamless, and for elbows, tees and tube fittings.

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Temper-embrittlement

A review of the literature

The phenomena of temper-embrittlement and the techniques used to evaluate embrittlement characteristics have been studied since the beginning of the century. This critical review of published work in the field is by Dr. T. P. Singh, chief metallurgist, Tinplate Company of India, Jamshedpur, and is based on the author's contribution in 'TISCO,' October, 1958

MOST OF THE ENGINEERING STEELS are used in a quenched and tempered state, the purpose of tempering being to induce a degree of softness and toughness in the steel. But it has been a well-known fact that many of the steels of the low-alloy type, after being quenched and tempered, lose their impact resistance and become brittle if after tempering the steels are cooled slowly or held for longer time in the temperature range 600—300°C. It is to this phenomenon that the name temper-brittleness has been given. A Ni-Cr steel cooled at different rates from the temperature of 650°C. gives a set of vastly different notched-bar impact values depending on the rates of cooling, as can be seen from Table I.

TABLE I *Effect of rates of cooling of quenched and tempered steel*

Method of cooling from 650°C.	Fracture energy, ft. lb. (Izod test at R.T.)
Furnace	7
Air	17
Oil	44
Water	55

As early as 1883, certain blacksmiths followed the practice of water-quenching the steels from the tempering temperature, to avoid lack of toughness, and this practice came to be known as water-annealing. Later on the desirability of quenching Ni-Cr steels after tempering was emphasized in several Krupp patents taken in the year 1900.[1] It was, however, during the First World War, when lots of alloy steels were used and needed for war efforts, that this problem of temper-brittleness became important. The notched-bar impact test became one of the important acceptance tests, and the steels had to be suitably heat-treated to give a high impact strength.

Comparison of the room temperature impact-test results was the basis of earlier investigators for studying the embrittlement in various steels after different heat treatments. The tendency of a particular steel to temper-embrittlement was expressed as the ratio of impact-test results after water-quenching and after furnace cooling respectively from 650°C., and this ratio was called the 'susceptibility ratio' of the steel.

Although a considerable amount of work has been done, and much experimental data were made available by the earlier workers like Dickenson,[2] Hatfield[4] and Greaves,[5] there were still wide divergencies in their results. Greaves and Jones,[9] however, discovered in 1925 that the addition of molybdenum to susceptible steels could reduce or remove the embrittlement, and this helped greatly to solve the practical aspect of the problem, but the cause for temper-brittleness still remained obscure.

Transition temperature

The limitations of and discrepancy in assessing temper-embrittlement of a steel by determining its susceptibility ratio became quite obvious as a result of Jolivet and Vidal's[3] work in the year 1944. They were the first to point out that the fracture energy of all ferritic steels varied with the temperature of testing, and that there was a rather steep drop in the impact energy of such steels at a relatively lower temperature, irrespective of the fact whether the steel was temper-embrittled or not. This marked drop in the fracture-energy/testing-temperature curve came to be known as transition temperature, and its exact location with respect to the curve will be discussed later on. Jolivet and Vidal further indicated that, as a result of temper-embrittlement in a steel, the portion of its fracture-energy/testing-temperature curve, including the transition temperature, was only displaced to higher temperatures as shown in fig. 1. It is quite clear from this

figure that, for the same steel, a set of three entirely different susceptibility ratios can be obtained depending on the temperature of testing. If temperature T_2 happens to be the testing temperature, then the steel would appear to be highly susceptible, while if the testing temperatures are T_3 and T_1, the steel would be only slightly or little susceptible respectively.

This new concept of the effect of temper-embrittlement has made it necessary to re-examine the earlier work, and has provided a new basis for the study of temper-brittleness. Even before Jolivet and Vidal, both Monypenny,[10] and Greaves and Jones[11] had found the variation of notched-bar impact results with the testing temperature, but had failed to realize its real significance. Since 1944, however, the degree of displacement of the transition temperature in the fracture-energy/testing-temperature curve became a measure of the amount of embrittlement.

As stated already, the point of inflection in the fracture-energy/testing-temperature curve is generally taken as the transition temperature.[8] There are, however, other methods employed by various workers for locating the position of the fracture-energy curves for purposes of comparison. Principles based on the nature of fracture, and also the amount of fracture energy, have been used for fixing the transition temperature in the curve. They are mainly as follows:

(i) The lowest possible temperature at which the fracture is 100% ductile.

(ii) The temperature at which fracture shows 50% brittle fracture.

(iii) The temperature at which fracture energy is reduced to some arbitrary value *e.g.* 15 or 20 ft lb.

(iv) The temperature at which fracture energy is reduced to half the value required for completely ductile fracture.

The amount of displacement of the transition temperature due to temper-embrittlement is taken to be a direct measure of the amount of temper-brittleness, although no reasons have been given to show that such a simple relationship exists.

The choice of any of the definitions of transition temperature does not matter greatly if the same definition is adhered to in any particular investigation. But if the shape of curves for various steels are basically different, then the choice of a particular definition may well matter.

The inflection in the curve (fig. 1) very closely corresponds to the presence of initial traces of brittle fracture in the test-piece. The percentage of brittle fracture progressively increases, as one comes down the slope of the curve till it is almost 100% brittle where the curve again flattens out at lower energy levels. Since this transition from ductile to brittle fracture takes place in both unembrittled as well as embrittled steels, it is of interest to see if the brittle fractures in them differ from each other in any respect. This aspect of fractography will be discussed later.

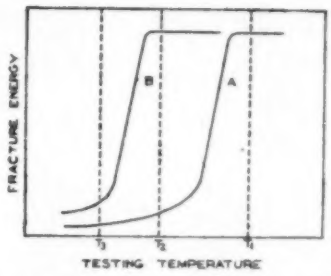
Metallographic evidence

Structures of specimens with and without temper-embrittlement have been found to show little or no difference when treated by normal etching reagents like picral, nital and even special carbide reagents.[7, 9, 18, 19] Cohen and others,[18] however, were the first to distinguish between an unembrittled and an embrittled specimen under a microscope, using an ethereal solution of picric acid mixed with zephiran chloride (zephiran chloride is a mixture of alkyl dimethyl-benzyl-ammonium chlorides). There was a marked grain-boundary attack in the case of the embrittled specimen while the unembrittled specimen showed just the normal etch without any suggestion of grain delineation. The other etching reagents mentioned above failed to produce any grain attack. The close co-relationship between the grain-boundary attack and temper-brittleness in the specimens examined by them was established, and they claimed to distinguish between unembrittled and embrittled specimens of a Ni-Cr steel when the difference in their transition temperatures was only 20°C.

It was discovered[19] later on that several other reagents, with picric acid as one of their constituents, also produced the grain-boundary attack, and in fact Woodfine[35] showed that aqueous picric acid was equally good in this respect.[34]

McLean and Northcott[19] suggested that the grain-boundary attack in the embrittled specimens revealed the original austenite grain boundaries and that the sub-boundaries observed by them inside a parent grain were due to the difference in orientations within the grain. There were no experimental proofs given by these investigators to support their assumption that the grain boundaries revealed by their etchant were the prior austenite grain boundaries.

Woodfine,[35] in an attempt to establish some relationship between the grain delineation and the prior austenite grains of an embrittled specimen, devised an experiment in which he thermally etched the prior austenite grains of specimen and subsequently temper-embrittled the same specimen to examine the grain-boundary attack. His technique consisted of thermally etching an electropolished Ni-Cr steel specimen by heating it for 1 h. at 900°C. and then quickly cooling it in the furnace tube itself under reduced pressure. The cooling from 900°C. to room temperature resulted in a bainitic structure. He drew a number of small rings on the polished and thermally etched surface of the specimen to help locate representative areas and took a number



1 *Fracture-energy/testing-temperature curves for a steel with (A) and without (B) temper-brittleness*

of photographs before and after a light picral etch. The specimen was then given an embrittling treatment by heating it to 650°C. for 1 h. under reduced pressure and allowed to cool slowly in the furnace. After removing all traces of the previous etch by light polishing and picral etch, the specimen was finally lightly polished and etched in aqueous picric acid and the representative areas were then rephotographed. The micrographs quite unmistakably showed that the grains that were thermally etched, *i.e.* which were prior austenite grains, were also the ones that were attacked by the aqueous picric acid after the embrittling treatment. The fact that the embrittling treatment was given to a steel with bainitic structure instead of the usual tempered martensitic structure need not, according to Woodfine, mean that the same results would not apply to the latter, *i.e.* the tempered martensite structure. This seems to be a reasonable assumption.

There were, however, some grains found after embrittling treatment in his other experiments which did not conform to prior austenite grains, and Woodfine thought that these sub-grains corresponded to ferrite grain boundaries.

Grain-boundary etch

The cause of the grain delineation in the case of the temper-embrittled specimens has been studied by a number of workers. Maloof,[17] took a few electron micrographs of a Ni-Cr steel with and without temper-brittleness, using zephiran chloride as the etching reagent. The embrittled specimen showed a clear grain-boundary attack, but Maloof could not be certain if the etching reagent had produced a groove or a ridge at the grain boundary. Woodfine[35] also studied the grain-boundary attack due to aqueous picric acid in temper-embrittled specimens with the help of an electron microscope, using negative formvar replicas. The grain boundaries when etched by aqueous picric acid appear to be dark at low magnifications. But when seen under the high magnification of an electron microscope, these were found to be distinct grooves. Woodfine did not find any evidence of any precipitate along the grain boundary except for the occasional carbide particles whose distribution was quite random in the matrix without any special preference for the grain boundary. He further observed that by increasing the etching time the grooves became deeper and wider. He used this observation to support his contention that the etching was not due to the presence of a film or particles of a distinct phase which would have a defined boundary with respect to the matrix, and would, therefore, have limited the depth of the groove to these boundaries; this is contrary to what has been observed.

Nature of fracture

A marked difference is noticeable in the brittle fractures of an unembrittled and embrittled steel even on visual examination. Suggestions were made quite early that the brittle fractures in an embrittled steel were always intergranular since, in the case of unembrittled steels, cleavage fractures[6, 20, 21] were observed. There was another school of thought,[4, 9, 22, 23] however, which would not entirely agree to this suggestion, and it remained inconclusive till McLean and Northcott observed in 1947 that the brittle fracture in a temper-embrittled Ni-Cr steel was intergranular. Unfortunately, they did not examine a corresponding brittle fracture in the same steel without temper-embrittlement.

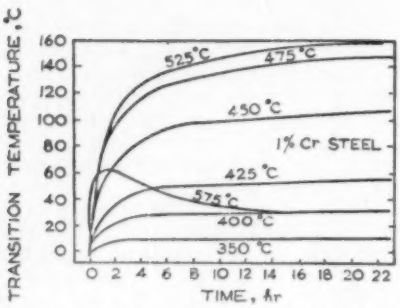
Entwisle and Smith[24] showed clearly with the help of two fractographs that in the case of a Ni-Cr steel, which was broken at –180°C. to give brittle fracture without temper-embrittlement, the fracture was distinctly cleavage, while the same steel with temper-embrittlement broken at the same temperature showed intergranular fracture.

Jaffe and Buffum[36] have recently shown by carefully preparing a cross-section at right-angles to brittle fracture in an isothermally temper-embrittled steel that the path of fracture mostly follows the austenite grain boundary.

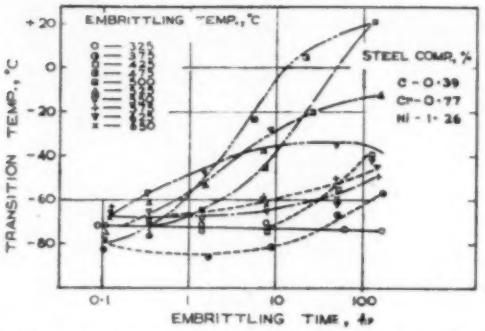
Woodfine,[35] using a similar technique, had found the same results earlier. In fact, based on the experiences of a number of investigators, the absence of intergranular fracture in a specimen which has failed in a brittle manner has come to be regarded as a definite proof that temper-embrittlement has not taken place in the steel while its (intergranular fracture) presence meant that the steel has developed temper-brittleness.

Effects of temper-embrittlement

Results of careful investigations have failed to show any apparent difference in the mechanical properties of several susceptible steels before and



2 *Transition temperature for a Cr steel plotted against time of treatment at various temperatures indicated on curves. The steel was originally quenched and tempered at 650°C.*



3 *Transition temperature v. embrittling time for various embrittling temperatures*

after the temper-embrittling treatment. In these investigations due precautions were taken to ensure that the steel with and without temper-brittleness had a comparable amount of tempering. This was ensured by taking two specimens which were quenched and tempered at 650°C. One of the specimens was then quenched from the tempering temperature and the other was allowed to cool slowly in the furnace from 650°C. Both the specimens were again reheated to 650°C. and this time the specimen which was cooled slowly in furnace before was quenched from 650°C. and the other one which was previously quenched from 650°C. was allowed to cool slowly in the furnace from the same temperature. Test results on these specimens showed no difference in their fatigue properties, hardness, yield stress, maximum stress or elongation.[5, 12, 13] So temper-brittleness does not seem to affect these properties. Woodfine,[35] however, has found in his low-temperature (–196°C.) tensile tests that the Ni-Cr steel *K* (2·92% Ni, 0·87% Cr, 0·33% C), after having been embrittled for 32 h. at 500°C., showed a decrease in the reduction in area from 57% to 35% and in the maximum stress from 96·5 to 84·1 tons/sq. in. This is attributed to the lowering of the brittle-fracture strength, as distinct from the ductile strength, due to temper-brittleness. In severely temper-embrittled specimens, slight decrease in the reduction in area at room temperature tensile tests has been observed in a few steels. The effect of temper-brittleness on notched bars, whether in impact or in slow-bending tests, is the same and is shown in fig. 1. Magnetic and electrical properties of specimens with and without temper-brittleness showed no significant difference.[9, 14]

X-ray investigations were also made using a back-reflection technique by Weill[15] as well as by Jacquet[16] and others on specimens with and without temper-brittleness. In the tough condition, Weill found that the (211) *Ka* doublet was resolved, while in the brittle condition it was not. Jacquet and others got entirely opposite results to this on a chromium steel. So nothing conclusive came out of this and the effects observed were probably due to the difference in ferrite grain size and carbide particles rather than to factors connected with temper-brittleness.

Isothermal kinetics

It has been generally found that specimens of susceptible steels held or allowed to cool slowly in the temperature range 600—300°C. develop temper-brittleness in a small or greater degree depending on several factors. The most important of these seem to be the temperature and time of holding in this range in addition to the inherent susceptibility of the steel and its previous history.

The effect of temper-brittleness which takes place in this range is completely removed by heating the specimen to above 650°C. and then quenching it in water to any temperature below 300°C.

The study of the separate effects of temperature and time is best done with the help of an isothermal embrittlement diagram where data obtained from isothermally treated specimens are plotted. Greaves and Jones[9] first studied the progress of temper-brittleness in isothermally treated specimens, but they used the room temperature impact results. They found, however, that the maximum temperature at which temper-brittleness developed was 600°C.

Vidal followed[25] the development of temper-brittleness with time at various temperatures by determining the displacement of the transition temperature. The steel studied had the following composition: 0·25% C, 0·30% Mn, 1·38% Cr. The specimens were austenitized and quenched to martensite and then tempered at 650°C. for 1 h. and water-quenched; they were then treated for

various times at different temperatures from which they were quenched in water again. The results are shown in fig. 2[25]

The curve for 525°C. shows the maximum embrittlement for long periods of tempering. The rate of embrittlement is quite high in the initial stages, which seems to slow down after some time and tends towards a steady value at longer times of holding. This feature seems to apply to curves at lower temperatures also. Only the initial rate of embrittlement and the total embrittlement tend to decrease with decreasing temperatures. The curve for 575°C. shows that the maximum rate of initial embrittlement attains the maximum value in about 2-h. time and then the embrittlement seems to decrease with increasing time.

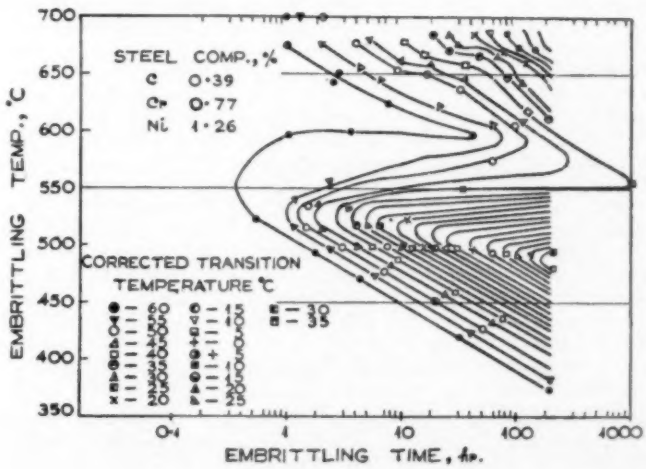
Jaffe and Buffum,[26] using a similar technique as employed by Vidal, studied temper-embrittlement in SAE 3140 Ni-Cr steel (0·39% C, 0·77% Cr, 1·26% Ni, 0·30% Si, 0·79% Mn), the results of which are shown in fig. 3. Their results differed in certain respects with those of Vidal's, but this was probably because of the difference in the compositions of their steels. Jaffe and Buffum, however, also observed that at 550°C. the embrittlement started decreasing after 48 h. This effect is somehow not indicated in their curves but is mentioned in the text.

Vidal, in the case of his steel, thought it to be a phenomenon similar to over-ageing in age-hardening steels. Jaffe and Buffum attributed this behaviour to the softening that might have occurred in the steel due to retempering at this temperature.

Woodfine[35] found similar effects in his steel *K* at temperatures of 550 and 575°C. At 550°C. the maximum embrittlement reached was approximately 95°C. in 6 h. and, on further tempering, the embrittlement slowly decreased to 74°C. after 128 h. Again, at 575°C., the embrittlement reached a maximum of about 72°C. after only ½ h. and then decreased to 39°C. after 128 h. Due to long tempering at these temperatures, a significant decrease in hardness had occurred. To investigate the effect of hardness on transition temperatures in steel *K*, Woodfine oil-quenched a group of bars from 850°C. and tempered them at 650°C. for 5 h., and he found on comparing the results for 1 h. and 5 h. tempering at 650°C. that a decrease of 27 V.P.N. numbers decreased the transition temperature by 8°C. At 550°C., the transition temperature decreased by 21°C. between 8 h. and 128 h., whilst the hardness fell by only 10 V.P.N. numbers; at 575°C. the transition temperature decreased by 33°C. between ½ h. and 128 h., whilst the hardness fell by 18 V.P.N. numbers. Woodfine thus observed that, provided the change in the transition temperature produced by a given decrease in hardness was independent of the actual value of the transition temperature, it would appear that the softening at 550 and 575°C. was not sufficient to produce the observed decrease in the temper-embrittlement at these temperatures.

He further went on to suggest that the decrease in the temper-embrittlement that occurred in steel *K* at 500 and 575°C. after long-time treatments did not appear to be due either to softening or the spheroidization of the grain-boundary phase, but was possibly the result of diffusion of chromium from the matrix into the carbide particles.

The maximum temperature at which temper-brittleness has been known to occur has been reported differently by different investigators, which



4 *Time-temperature diagram for isothermal embrittlement of Ni-Cr steel. All specimens previously austenitized for* 1 *h. at* 900°C. *and water-quenched. Specimens treated at* 660°C. *and below were also given a prior temper of* 1 *h. at* 675°C. *and water-quenched; plotted times for* 675 *and* 700°C. *are adjusted accordingly*

it at least partly due to the different steels used by them. Woodfine[35] found that in steel *K* temper-embrittlement could occur at and above 600°C., but the magnitude of the effect was small.

Recently Jaffe and Buffum,[36] on the other hand, have reported that SAE 3140 steel showed a very rapid embrittlement at 675°C. In fact, they extended the isothermal study of temper-brittleness to temperatures up to 700°C. (690°C. being the Ae_1 of the steel) and found the presence of the upper nose in the time-temperature diagram for isothermal embrittlement as shown in fig. 4. The upper nose, as seen in the diagram, corresponds to 675°C. and, according to the authors, the upper nose embrittlement was most severe at this temperature. There is also a deep bay or region of slow embrittlement just above 600°C.

Specimens embrittled in the upper nose region, however, showed poor response to ethereal picric acid etch. Also they showed fractures quite different to the ones embrittled at lower temperatures, which was apparent even to the naked eye.

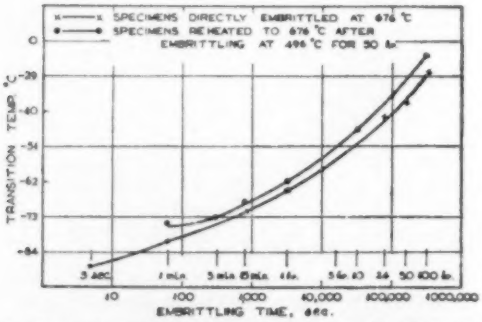
Bhat and Libsch,[37] using the same SAE 3140 steel, studied the effects of low-temperature embrittlement on the subsequent upper nose embrittlement and *vice versa*. Their results appear in figs. 5 and 6. It can be seen from the diagrams that the effect of lower nose embrittlement on the subsequent upper nose embrittlement is almost negligible. But if the lower nose embrittlement follows the upper nose embrittlement, as shown in fig. 7, the specimen fails to attain values which it would otherwise have done if the latter was absent.

Woodfine[35] studied the effect of previous treatment on the temper-embrittlement produced in a subsequent treatment at different temperatures in the case of the Ni-Cr steel *K* (2·92% Ni, 0·33% C, 0·87% Cr). He also found that, if both the successive treatments were at temperatures in the lower region (550°C. and below), the effect produced was as if only the last treatment was given and the first one omitted. In the case of the specimen which was previously treated at 575°C. and above and then subsequently treated at a lower temperature of 500°C., the amount of embrittlement was less than if the specimen were treated at 500°C. alone for the same length of time—a result similar to that of Bhat and Libsch.

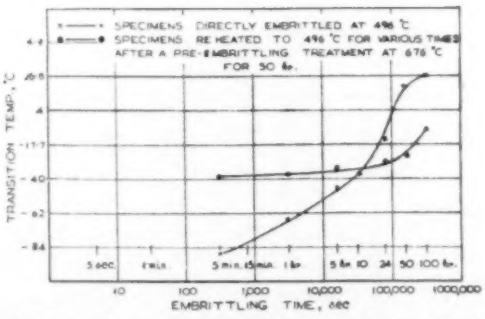
From the kinetics of the temper-brittleness reaction studied at various temperatures (fig. 5) there seems to be a strong suggestion that two different mechanisms are operating: one above the 600°C. responsible for the upper nose reaction and the other between 600 and 300°C. responsible for the lower region embrittlement. This is also borne out by the metallographic and the fractographic evidence available.

Effect of previous treatment in the upper nose region (670°C.) on the subsequent treatment in the lower temperature region (500°C.), as pointed out above, has been taken to mean that the embrittlement taking place at the lower region[37] is sensitive to structure, *i.e.* a structure containing well-developed ferrite grain boundaries appears less susceptible to embrittlement in the lower nose region.

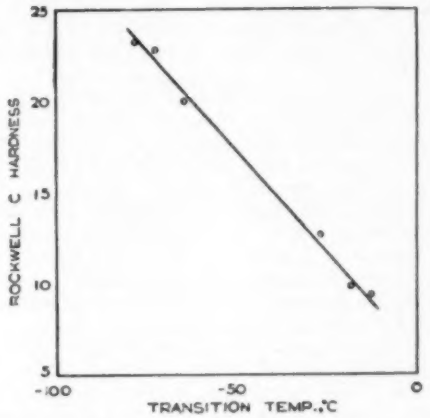
On the other hand, the effect of embrittlement at 500°C. is completely removed by reheating to 670°C. for very short times, and has no significant influence on the subsequent development of embrittlement at 670°C. This behaviour is suggestive of either nucleation and growth of a precipitate and its solution with increased solubility at higher temperatures or segregation of solute elements to prior austenite grain boundaries[33] and again a decrease in the grain-boundary segregation at higher temperatures due to the increased thermal mobility of carbon atoms, and also as suggested by Woodfine, as the ferrite lattice expands, from a decrease in the lattice strain produced by the presence of the carbon.


SPECIMENS DIRECTLY EMBRITTLED AT 676 °C
SPECIMENS REHEATED TO 676 °C AFTER EMBRITTLING AT 496 °C FOR 50 h.
TRANSITION TEMP, °C
EMBRITTLING TIME, sec.

5 *Comparison of embrittlement in specimens directly heated at 676°C. with specimens reheated to 676°C. after prior embrittlement at 496°C. for 50 h.*


SPECIMENS DIRECTLY EMBRITTLED AT 496 °C
SPECIMENS REHEATED TO 496 °C FOR VARIOUS TIMES AFTER A PRE-EMBRITTLING TREATMENT AT 676 °C FOR 50 h.
TRANSITION TEMP, °C
EMBRITTLING TIME, sec

6 *Comparison of embrittlement in specimens directly embrittled at 496°C. with specimens reheated to 496°C. after prior embrittlement of 676°C. for 50 h.*



7 *Relationship between Rockwell hardness and transition temperature of a Ni-Cr steel tempered at 675°C. for increasing length of time*

The segregation of solute atoms at the grain boundary rather than the nucleation and growth of a new phase appears to be the possible cause for the lower temperature region embrittlement. Electron microscopic study has failed to show the presence of any new phase[35] at the grain boundaries of embrittled specimens and this seems to rule out the nucleation and growth mechanism of a new phase as a possible cause for temper-brittleness.

As for the embrittlement taking place at 670°C., Bhat and Libsch suggest that it may be related to a permanent structural change which is continuous with temperature and time. They further point out that this structural change is consistent with the development of well-defined ferrite grains at 670°C. and their growth with increasing time. Woodfine,[34] who also studied Jaffe and Buffum's earlier[26] result, made the same observations that, in the absence of any metallographic or fractographic evidence, the increase in the transition temperature at 670°C. was not due to temper-brittlement but probably to the increase in the ferrite grain size and carbide coalescence with the increase in temperature. He also showed, using Jaffe and Buffum's results (fig. 7),[34] that there was a significant correlation between the transition temperatures and the Rockwell hardness values for the specimen treated at 670°C. for increasing length of time.

Jaffe and Buffum[26] maintained that the upper nose embrittlement could not be due to ferrite grain growth and pointed out that an increase of more than 90°C. in the corrected transition temperature (correction for fall in hardness) was observed during the upper nose embrittlement with no noticeable change in ferrite grain size. They, however, did not suggest any alternative mechanism to explain the upper nose embrittlement.

Temper-embrittlement in various types of steels

From the practical aspect of the problem, it is the Ni-Cr steels of the low-alloy type which are considered important so far as the susceptibility to temper-brittleness is concerned. This is, first, because these steels are very widely used and, secondly, they seem to be very prone to temper-brittleness.

Steels containing Ni, Cr and Mn or even any two of these elements form the largest group of susceptible steels.

Plain nickel steels are not usually considered to be temper-brittle.[3, 4, 8] Woodfine,[35] however, found that a 3% Ni steel showed a very slight embrittlement when treated at 500°C. for 24 h. The actual displacement of the transition temperature was only 7°C., and there was also a slight but definite grain-boundary etch when treated with aqueous picric acid.

Addition of phosphorus to a nickel steel was found to increase the susceptibility of the steel considerably.[35] In a steel with 0·1% P and 2·94% Ni, the transition temperature was found to increase by 84°C. as compared to 7°C. only in the case of 3% Ni steel with 0·01% P. The properties of steel with 0·1% P were adversely affected even in the absence of temper-brittleness. Compared to the steel with 0·01% P, the maximum fracture energy of the high-phosphorus steel (0·10% P) decreased from 72 to 52 ft. lb. and the transition temperature was raised from −34 to +12°C.

Plain chromium steels are also susceptible to temper-brittleness, and Jolivet and Vidal[3] have shown that the temper-embrittlement produced by a given treatment increases with increasing chromium. Similar results have been found with respect to manganese also.[28, 29]

The effect of nickel, chromium and manganese in combination becomes complicated. Increasing the amount of any one of the elements when all the three are present in a steel increases the susceptibility.[12, 28] Nickel alone has a very slight effect on the susceptibility, but when present in a susceptible steel, its (Ni) increase also increases the susceptibility. Effects due to manganese and nickel are increased by the addition of chromium, but increasing the chromium content of a steel containing manganese and nickel has an effect which is quite independent of these elements.

Molybdenum is exceptional in its effect and, when present in amounts up to about 0·4%, it decreases the susceptibility of otherwise susceptible steels.[30] But when present in larger amounts (2%) it has been found to make the steel susceptible.[25]

Vanadium and tungsten have both been found to increase the susceptibility of steels.[3, 25]

There have been suggestions that temper-brittleness was due to the precipitation of iron nitride.[31] Woodfine,[35] using steels containing titanium, found that its presence did not help in preventing temper-embrittlement in susceptible steel. Since titanium is a strong nitride-forming element, he ruled out the possibility of the precipitation of iron nitride as a cause for temper-brittleness. The addition of 0·5% Al to a chromium steel did not affect the susceptibility.[3]

Effect of microstructure and grain size

Development of temper-brittleness in steels with structures other than tempered martensite has also been studied by a few workers. Pellini and Queneau[27] obtained fracture-energy curves for specimens temper-embrittled after various isothermal treatments. Temper-embrittlement in their pearlitic specimens was less than in the tempered martensitic specimens after the same treatment.

Woodfine's[35] results also show that temper-brittleness can occur in pearlitic and bainitic micro-structures as well as in tempered martensite. For the particular steel *K*, he found that, for the same embrittling treatment, the amount of embrittlement $\Delta\theta$ ($\Delta\theta$ being the rise in transition temperature) was 196°C. for tempered martensite, 110°C. for bainite and 51°C. for pearlite. All these embrittled specimens showed heavy grain-boundary attack when etched with aqueous picric acid.

An increase in the austenite grain size seemed to increase the susceptibility. Hurlich[32] and Woodfine[35] both found that the total embrittlement after a given time as well as the rate of embrittlement was much greater than for the same steel in the fine-grained condition.

Plain carbon steels

Temper-brittleness in plain carbon steels has become a thing more of academical interest than anything else. Plain carbon steels with less than 0·5% Mn are usually not considered to be susceptible. Jaffe and Buffum, and Bhat and Libsch, however, made a suggestion that these were so extremely susceptible that even quenching them from above 600°C. did not prevent embrittlement. In other words, all plain carbon steels suffer from temper-brittleness even though they have escaped normal embrittling treatment.

Jaffe and Buffum argued that 'the hypothesis that plain carbon steels develop temper-brittleness so rapidly that they become embrittled even when quenched from tempering temperature then leads to the predictions that (i) without the deliberate embrittling treatment the transition temperature of the carbon steel is higher than that of the alloy steel, and (ii) with the embrittling treatment the transition temperatures of the two steels are roughly equal.'

They had taken precautions to ensure that the plain carbon and the Ni-Cr steel used had comparable hardness and grain size. The transition temperature was taken as the temperature at which the fracture was 50% brittle. Employing this definition of the transition temperature, they found a difference of 30°C. in the transition temperatures of plain carbon steel and alloy steel without the embrittling treatment. After the embrittling treatment, the difference in the transition temperatures of the two steels became 5°C.

Woodfine[34] re-examined their results and showed that, if the transition temperature is defined as the lowest testing temperature at which the fracture is 100% ductile (as used by Jaffe and Buffum in their other investigations), the following results are obtained: (1) the difference in the transition temperatures of the two steels without embrittlement became only 5°C. instead of 30°C., as found by Jaffe and Buffum, and (2) in the case of steels after embrittling treatment the difference in the transition temperatures became 30°C. instead of 5°C., which was exactly the reverse of what Jaffe and Buffum had found.

Jaffe and Buffum[36] have recently suggested that the temper-brittleness occurring in plain carbon steels is probably of the upper nose type.

Whether or not temper-brittleness does take place in plain carbon steels, it seems almost certain that it does not show characteristic features of the normal temper-brittleness.

Entwisle and Smith[24] found that a 0·4% C-0·6% Mn steel treated at 480°C. for 24 h. showed only 5% intergranular fracture when broken at −180°C., and they concluded that the steel was not susceptible to temper-brittleness in the normal sense of the term. The supposedly temper-embrittled specimens of plain carbon steels did not show the characteristic grain-boundary attack when etched with zephiran chloride.

Conclusions

It appears that the upper nose embrittlement reported by Jaffe and Buffum[36] in the SAE 3140 steel and the embrittlement taking place in the same steel at relatively lower temperatures, *e.g.* 500°C., arise from two different causes. Embrittlement taking place at lower temperature (600°C. and below) corresponds to the normal temper-brittleness as generally understood. It is important that, in a future study of temper-brittleness, the upper nose embrittlement should not be confused with temper-brittleness.

Specimens with temper-brittleness show an inter-

granular fracture when broken in an impact test. They also show a marked grain-boundary attack when etched with reagents based on picric acid and containing small traces of surface-active compounds.

Temper-brittleness displaces the fracture-energy/testing-temperature curve to higher temperatures, but its presence is not revealed by any of the normal mechanical tests. Physical properties of susceptible steels are also not affected by temper-brittleness. Plain carbon steels with less than 0·6% Mn do not show any temper-brittleness.

Steels containing Mn, Cr, W, Mo and Ni show susceptibility to temper-brittleness. Molybdenum when present up to a certain percentage (0·4%) decreases the susceptibility. Nickel when present alone does not usually make the steel susceptible, but in combination with high phosphorus, it has been found to increase the susceptibility.

Temper-brittleness has been found to occur in susceptible steels with tempered martensitic structure as well as with any other directly transformed product of austenite, *e.g.* pearlite, bainite, etc. The amount of temper-brittleness occurring in steel with tempered martenisitic structure is, of course, much more than in the same steel with pearlitic or bainitic microstructure.

The theory most favoured to explain temper-brittleness is the segregation of solute atoms at the grain boundaries. No evidence has been found of the presence of any precipitated phase at the grain boundaries.

The upper nose embrittlement occurring at 675°C. is possibly due to the ferrite grain growth and coalescing of carbide particles.

Long-time tempering at higher temperatures (650°C.) has been found to decrease the susceptibility of steels treated subsequently in the embrittling range. This suggests a practical possibility of minimizing the danger of temper-brittleness in susceptible steels.

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Tantalum

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An example in which the high corrosion resistance of tantalum is essential is as 'biologically acceptable' metal in surgical repair work. Implants of tantalum gauze are adopted for cases where muscular or soft tissue is missing or has been removed. Tissue grows over and incorporates the fine gauze, the tantalum being a bridge for such tissue growth. Fine tubes of very thin tantalum foil are used to encase severed nerves, while in plastic surgery tantalum wool is used to pad areas of missing flesh and tiny lengths of tubing for repairing small bones. Some thousands of skull injuries have been repaired with tantalum plates, while miles of wire have been applied for surgical sutures.

Two other applications of tantalum are to be found in the electrical industries. Tantalum and lead plates in dilute sulphuric acid constitute a rectifier or trickle-charger, for when a.c. is applied, current will flow from the lead to the tantalum and not in the opposite direction. The fact that tantalum forms an anodic film of unusual stability is also applied in capacitors incorporating one tantalum electrode and one of silver in an electrolyte, the tantalum element being always positive in the circuit. Here again the tantalum workability down to 0·0005 inch thick for such elements is of advantage.

In radio valves and electronic tubes the high melting point of tantalum and its strength are points further added to by the metal's 'getting' properties for trace of residual gases in the tubes. By taking the anodes and grids to 2,000°C. during processing, any gases present are evolved and the metal subsequently absorbs any gas within the valve or tube.

This diversity of applications of tantalum illustrate how an intractable metal first used for lamp filaments has outbid even tungsten in finding applications in the industrial field.

Flat-rolled steel

Development as a structural element

In recent years, the proportion of flat-rolled steel turned out of the world's rolling mills has steadily increased everywhere. In the United States of America in particular, this proportion is very high: up to 65% of the total. This development, which is by no means at an end, has been promoted by the fact that an appreciably larger range of structural elements in all branches of industry hitherto cast, or made of angles and other sections, is now being fabricated from steel plate and sheet. In the following article, published in 'Lerhen und Leisten,' 2058, p. 271, by Guenther Frank, this development is discussed in detail and the advantages of the new techniques outlined

FORMERLY, FLAT-ROLLED STEEL was used, as suitable, for covering large areas or for making hardware and packaging goods. Alternatively, it was used for motor vehicle bodies. Designers had not attempted to use it also for load-carrying elements. The use of flat-rolled material for this purpose required a considerable modification of current design practice and technology. A new set of static conditions faced the engineer with the problem of finding new bases and principles of calculation and thus of devising a new elementary theory which would satisfy the requirements of flat-rolled metal as the new structural material.

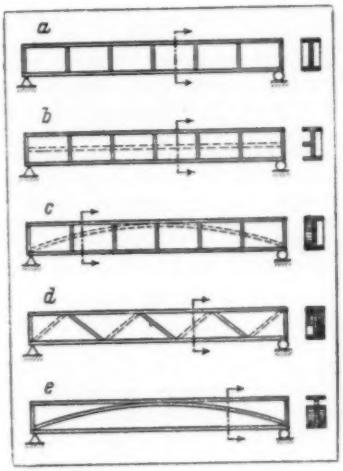
There are, as yet, no uniform and generally valid standards for shaped plate- and sheet-metal elements, but the first steps in this direction have been taken and it is the particular achievement of the Forschungsgesellschaft für Blechverarbeitung in Düsseldorf to have given guidance and leadership in this respect. Flat-rolled steel is now being widely used not only in the motor industry, and in the construction of ships and containers (tanks), but also in the construction of machinery and plant generally and in the electrical industries.

Replacement of rolled sections

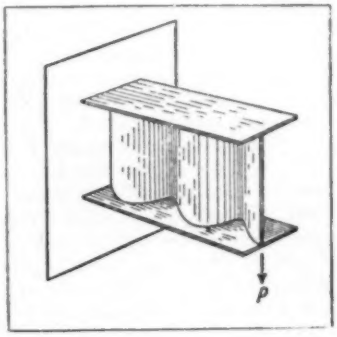
Frame members of flat-rolled steel now replace rolled sections which have for decades dominated the field of structural engineering. Stressed-skin construction in motor bodies enables the vehicles to be built much lighter and faster and the casings of many machines now fulfil the function of a load-carrying element. The promotion of any considerable industrial expansion depends on the economy of mass production in long runs, saving in weight, and thus, lower material costs. Flat-rolled steel construction satisfies all these requirements, and it is no accident that generations of engineers and technologists had to struggle with the problem of lightweight construction, before a broad basis for its general adoption on an industrial scale could be created. In this context, lightweight construction simply means saving weight by correct shaping and arrangement of the structural material, without sacrificing rigidity and strength to considerations of weight.

Sheet, strip and plate material of all thicknesses and sizes are now obtainable for all uses: from the ordinary commercial low-carbon steels through the range of high-strength carbon steels to the special alloy materials. These steels, depending on the alloying constituent(s), can be used for purposes requiring high resistance to corrosion and acid attack, or for resistance to high temperatures in boilers, plant or pipelines. Flat-rolled steel can also be supplied with various types of surface finish, ranging from tinned strip or tinplate to the special nickel-clad or chromium-plated tank plates. In this connection, mention may be made of the plastic-clad sheets recently put on the market, which are particularly resistant to acids and alkalis and have a high degree of ductility. In addition, they can be supplied in very attractive patterns and colours which, as a true 'finish,' saves any further surface treatment of the completed work.

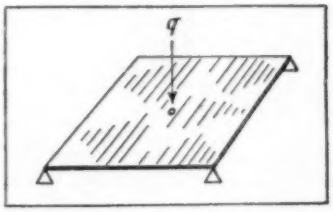
Improvements in welding technique have played a great part in this development, since it is one of the pre-requisites of successful design that not only should the individual elements thereof be well-conceived but that they should also be assembled and joined together by rational and technically sound methods. It is not so very long ago since



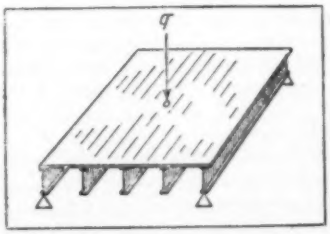
1 *Flat metal stiffeners on solid web girder*



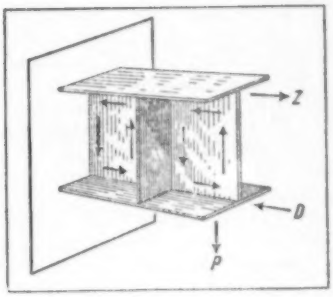
3 *Two-dimensional web design*



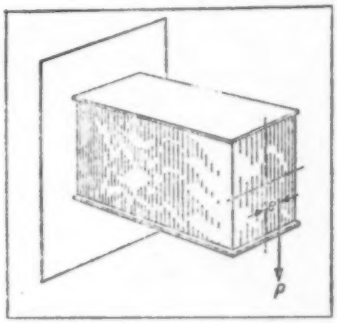
5 *Simple design for low loading and uniform distribution of loading forces*



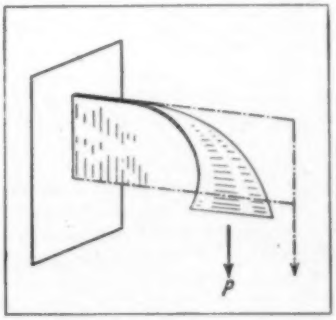
7 *Fitting longitudinal (I-) stiffeners to obtain increased bending rigidity*



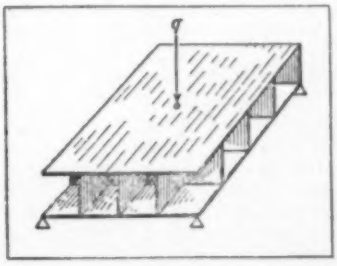
2 *Thin-web I-beam with cross-stiffening against lateral loads*



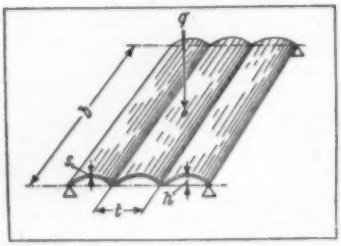
4 *Box-beam web*



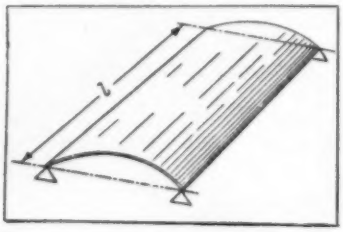
6 *Flat-rolled-metal construction, using deep, narrow girders for low, unidirectional loading*



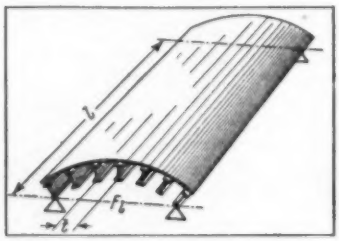
8 *Duplex (cellular) plate for greater torsional and bending rigidity*



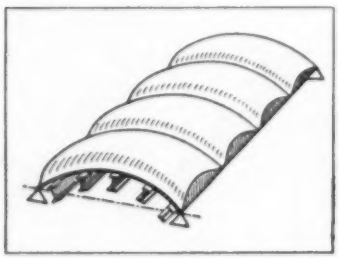
9 *Orthotropic, corrugated plate for higher shearing and torsional rigidity*



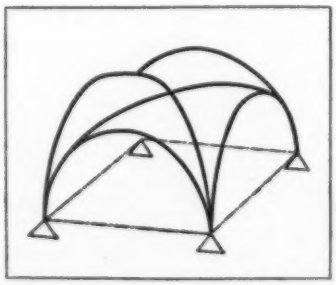
10 *Shell structure ensuring higher bending and torsional rigidity*



11 *Shell structure strengthened by fitting longitudinal stiffeners*



12 *Two-dimensional convex shell structure with longitudinal stiffeners*



13 *Three-dimensional shell structure*

bridge members were riveted or bolted together, the proportion of the connections to the total weight of the structure sometimes reaching 30%. The perfected welding methods of the present day have for the first time enabled the design of light and elegant, although structurally far stronger, flat-rolled steel constructions. Thus, strength and shapeliness can be combined, and if a riveted pre-war bridge—where one has survived—is compared with a welded structure, the contrast between the old and the new is almost as great as between Robert Fulton's first paddle steamer and a modern destroyer.

The 'natural' forms

It may be mentioned here that in the advance of engineering methods represented by flat-rolled steel construction, nature may have stood sponsor to a far greater extent than many engineers and designers may be willing to acknowledge. The first of these fundamental 'natural' forms is the skeleton (shell-frame) structure. This is a lattice framework of bars, either hinge-jointed together or forming rigidly-connected girders for truss frames. Since the individual structural members may be shaped as required a flat-rolled plate girder is far more easily adapted to the static conditions of any particular case. In addition, it provides ultimately a far more elegant and uniform structure than the section-bar design formerly used.

Geometrically, the elements of a skeleton structure are essentially slender bars which, as truss members, only have to experience longitudinal stresses or, as girders, additional bending stresses. As these structures increase in height and decrease in material thickness, they begin to merge in the next form of structure—plate construction.

The elements in this case are plates, *i.e.* walls, which are loaded only by forces acting in their own plane. The forces acting perpendicularly to the plane of the plate are absorbed by webs or 'diaphragms.' Such webs are found in precision instruments, in vehicle undercarriages, in ships' bottoms and in bridge decking. Such membranes obtained their stiffening from beaded depressions, or by welded-on stiffening ribs. This method is used particularly in shipbuilding.

If the torsional stresses are high, recourse is had to cellular construction. In this case, the cell walls act—depending on the nature of the load—either as plates or as diaphragms. If the edges of the resulting folded structure are infinitely closely approached, shell structures are obtained which are based on the realization that the sphere is the shape best able to support multilateral loading.

Spherical containers or tanks, buoys, valve floats, and gas bottles, are only a few examples of the

practical utilization of the principle of uniformly distributed loading.

While the designer has available a great variety of widely differing profile shapes, all the structural elements stemming from the three above-mentioned fundamental forms of sheet construction are based on the same design principles. If the material is concentrated as far away as possible from the section axis (centreline), particularly high moments of inertia are obtained. Consequently, deep, thin-walled, I and U sections are most suitable for high bending stresses. If the torsional load is considerable, thin-walled, plated box beams will be found suitable.

The difficulties encountered in regard to the load distribution at section joints and points of application of the loading forces cannot be surmounted by welding techniques alone, since it is just at the joints that stress concentrations and alternating stresses which may lead to hair cracks and fatigue failures are to be feared. To meet this danger, the closed sections are allowed gradually to merge in open forms, thus enabling deflection of the structure under torsional stress.

The strain-varnish testing method

The testing of complicated structures is often attended by considerable technical difficulties. To overcome these, Prof. Kloth and Dr. Ing. Bergmann, of the Institute for Fundamental Soil-Mechanical Research in Brunswick-Wolfenbüttel, have devised a strain-varnish method. This method enables the directions of principal stress to be determined, and the correct stress distribution thus to be obtained. For such a test, the flat-rolled metal part in question is coated with a brittle varnish. This cracks, when the component in question is loaded, perpendicularly to the first principal stress direction, and thus indicates the direction of stress.

Beading (corrugating) has already been mentioned. This is used for plate-stiffening. Beads or corrugations are produced by various methods —for instance, pressing or rolling—and sufficiently deep beads can also be produced by flanging.

The sheet sections are shaped either by flanging or pressing, as well as by drawing, on hydraulic or eccentric-driven presses. Stepwise deep-drawing calls for the use of certain deep-drawing steel grades, specially suited for this purpose. Economy of manufacture calls for long runs, since tooling costs are very high. In hydraulic deep-drawing, the position is different. In this case, in view of the uniform pressure distribution, particular material forms can be economically produced in a single operation, even in short runs. The peculiar characteristics of the spinning lathe limit its use to concentrically shaped parts.

Seam joints, in addition to the widely used welding method, can be made by riveting and, recently, by cementing. This probably gives the most uniform distribution possible of the loading forces. However, the technique of cementing (bonding) sheet steel structures is still in its infancy.

The fact that the rigidity per degree of elastic modulus of sheet metal structures is often twice as high as in the case of cast components has similarly revolutionized design thinking in electrical engineering. Nowadays, motor shields and casings, rotor spiders and bed plates are often made in cellular construction.

Stressed-skin construction

The stressed-skin form of construction in which the shell plating itself is made to act as a structural element belongs in its origins to the aircraft industry. Meantime, however, it has been adopted by the designers of rail and road vehicles and modified to suit the conditions encountered. The new corridor train coaches with stressed-skin body are less than 40% of the weight of their predecessors, in which the massive undercarriage, built of rolled sections, was the load-carrying member.

Components for lightweight construction, in the form of thin-walled, metal shells, are often filled with materials of lower specific gravity, rubber, polyvinyl chloride or polystyrene, for example. Nature was again the model, since plant stalks, bones and mushroom caps all have a soft, light filling material, the layers of hard enclosing fibres giving them support and strength.

Flat-rolled metal construction is also met with underground: conveyors, spiral chutes, trucks and steel arches are all principally of flat metal. Even worm conveyors, drill posts and crane jibs are increasingly departing from the former, massive form of construction and have become lighter, stronger and more elegant.

Since flat-rolled metal construction is on its home ground, in the case of tanks, containers, etc., it will be enough briefly to mention that here, too, in addition to the usual materials, unalloyed or alloyed, the steel-plastic synthesis is increasingly developing. Containers for the foodstuffs industry, foil-lined large containers for the chemical industry and non-corroding, centrifugally-sintered small parts for chemical engineering construction form an ever-growing field for the new materials. Conduits for hydroelectric power stations, steam mains, fan ducts, expansion chambers, expansion joints, and coiled tubing of spirally wound steel strip, are one group of flat-rolled steel construction, representing a considerable tonnage.

However, not only structural engineering, such as bridges, in which large spans without inter-

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Yttrium

YTTRIUM METAL, long considered too brittle for structural uses, has been transformed by Bureau of Mines, U.S. Department of the Interior, into a pliant, easily formed material that may be the answer to difficult problems in atomic-reactor and missile design.

The secret of changing hard-to-work yttrium into a metal so ductile it can be cold-rolled to thin foil was discovered by scientists at the Bureau's Albany, Oregon, laboratories. Through research in high-purity metals being conducted in co-operation with the Atomic Energy Commission, metallurgists at Albany recently learned that yttrium's forming qualities could be improved by ridding it of dissolved gases. The high-purity yttrium made by the Bureau contains only about 0·02% oxygen, which is the principal element limiting ductility.

Yttrium is found in close association with the rare-earth elements and, like the rare-earth metal ytterbium, takes its name from the village of Ytterby in Sweden, where some of these metallic elements were first obtained. Also, like the rare earths, yttrium is difficult to separate from its companion metals. It is produced commercially in relatively small amounts and is quite expensive.

High-purity process

Despite its high cost, the best yttrium metal available before the Bureau's recent attainment had limited ductility, even though it was considered almost free of contaminants. The technique for achieving the high purity required for ductility is similar to the Kroll process which the Bureau of Mines engineered from laboratory to commercial scale for titanium and zirconium. While this method does not reduce the costs of making yttrium metal, it will not increase them significantly, the Bureau said. Moreover, it makes possible production of a material with greatly improved ductility which should pave the way for wider use.

Inability to obtain yttrium in non-brittle form has prevented development of quantity uses thus far. Now that ductility has been attained—laboratory workers have reduced small ingots as much as 95% in thickness without annealing—yttrium's other properties suggest its use in building atomic reactors and atom-powered missiles.

One of yttrium's virtues in the nuclear field is its relatively low thermo-nuclear cross-section, which means it has less resistance than many other materials to the passage of neutrons needed to sustain a chain reaction. This is an important factor in rating construction materials for atomic reactors and engines. The metal also has a relatively high melting point of 2,825°F., roughly equivalent to that of carbon steel. This is high enough to enable yttrium to withstand the temperatures encountered by reactor parts that might utilize its thermo-nuclear properties.

Bureau of Mines' researchers foresee a potential value for yttrium in alloys, where its qualities could be blended with those of other metals to create new engineering materials to meet increasingly rigorous specifications. Moreover, they said, availability of this metal in ductile form will permit a re-evaluation of its possible applications in other important fields. Although only small amounts of high-purity yttrium are available as yet, enough has been obtained for limited studies of the metal's physical and chemical characteristics.

Yttrium is the latest of several metals obtained in very pure form at the Albany laboratories. Others are chromium, molybdenum, nickel, tungsten and vanadium.

Flat-rolled steel

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mediate supports have been made possible by the use of solid-web girders, has benefited by the adoption of sheet-steel construction; ordinary house-building has likewise learnt to use sheet and strip materials. Whole frontages are already being made of sheet steel and, by use of enamelling, very attractive effects have been obtained, with complete freedom from corrosion attack. For noise-suppression, the space between the thin sheet material forming the inner and outer surfaces is packed with some heat- and sound-insulating substance, since noise is a very disagreeable accompaniment of 'unsuppressed' sheet material. This trouble can be overcome by increasing the stiffness of the structure to the maximum limit, or by taking measure to suppress the noise and providing an insulating sheath. Sound can, however, also be absorbed by applying a coating of a suitable plastic on the inner side of noise-sensitive parts.

The advantages attained by resort to flat-rolled steel construction are so considerable that, at the present time, no progressive designer can afford to disregard them, since lightweight construction enables structural elements to be manufactured with a considerable saving in materials. Furthermore, the elements of a structure can, so to speak, be 'made to measure,' by drawing, pressing or flanging; and sheet material has a very suitable heat conductivity and is an excellent material for heat exchange. Moreover, structural forms can now unite efficiency with elegance—almost as is done by nature.

The care of platinum thermocouples

Platinum thermocouples are sensitive instruments. Exceptional care and control are given to their preparation in order to ensure reliability in service, but their accuracy and working life can be adversely affected if a few simple precautions are neglected. The following account summarizes the points made by Mr. H. E. Bennett, F.I.M., in 'Platinum Metals Review,' October, 1958

THE PLATINUM THERMOCOUPLE of the platinum : rhodium-platinum variety, is probably the most accurate and convenient method available for measuring high temperature in industrial processes, and if correct conditions of use are understood, the thermocouple with auxiliary equipment is both robust and reliable. Because of its small size and low thermal capacity, the couple can be introduced at the point of measurement and will respond immediately. The protective sheath can be small and inexpensive, and with compensating leads the indicating point can be arranged well away from the measuring point, which is often hot and dirty.

Because of their nobility, platinum and its alloys will withstand much more adverse conditions than base metal thermocouples, but they have their limitations. The melting point of platinum is 1,769°C. and that of its highest melting rhodium alloy useful for thermocouple purposes is about 1,890°C. That is, therefore, the absolute limit of temperature a platinum alloy couple will measure, and then only for a short time. In immersion pyrometry for measuring the temperature of molten steel, platinum alloy couples, in conjunction with a quick-response indicator, can be used momentarily at temperatures approaching these high values. For many other purposes thermocouples must be kept at temperature for long periods, and caution must be observed if the couple is to have a reasonable life. Grain growth, which occurs particularly in pure platinum when couples are kept at temperatures around 1,500 to 1,600°C., need not be deleterious if the wire is not under appreciable stress and is kept free from contamination, but these conditions are not easily attainable.

Effect of prolonged heating

The harmful effect of prolonged heating of the couple at high temperature is the migration of the alloying constituent from one wire to the other, the platinum element tending to pick up rhodium, the e.m.f. at any given temperature becoming progressively less. If the platinum wire is protected by a refractory insulator the rhodium pick-up is minimized.

It is difficult to give a definite figure for the expected useful life of a thermocouple as so much depends upon particular conditions. Instances have been recorded of platinum : rhodium-platinum couples being in use continuously for three years at about 1,300°C. with a loss in e.m.f. equivalent to only a few degrees. Obviously the higher the temperature of operation the greater the migration of rhodium and consequently the more rapid the change in calibration.

TABLE I *Pick-up of rhodium-platinum alloy compared with pure platinum*

Increase in e.m.f. at 1,200°C. against pure platinum by addition of 0·01% rhodium	Microvolts
Platinum	150
1% rhodium	20
5% „	11
10% „	6
13% „	4

It can be seen in Table I that a small pick-up of rhodium does not have such a great effect on a rhodium-platinum alloy as it does on pure platinum. For this reason a 1% rhodium-platinum : 13% rhodium-platinum thermocouple should be used where changes in calibration must be kept to a minimum over long periods. In practice, however, the thermocouple usually deteriorates by contamination from other sources before calibration has changed significantly.

Contamination by metallic vapours

Metallic vapours of lead, zinc and bismuth are among the commoner contaminants which readily alloy with platinum to form brittle intercrystalline constituents and so to cause premature failure of thermocouples. The presence of these impurities is often quite unsuspected, but they can arise from solders, galvanized iron, brass, or molten glass that may be present in the furnace, and couples need to be protected against them. The couple wires must be insulated from each other and the use of a double-bore refractory tube is recommended. This assembly should be placed in a closed refractory

sheath capable of withstanding the required temperature, as well as temperature changes and the furnace atmosphere. Silica may be used continuously at temperatures up to about 1,000°C. but for higher temperatures fireclay, graphite, silicon carbide mixtures or aluminous porcelain may be used. Fused alumina refractories are excellent for use in a hydrogen or other reducing atmosphere, in which platinum will be contaminated by any material containing silica, but they are not resistant to thermal shock and must be heated and cooled very slowly to prevent cracking. Sometimes for added protection it is advisable to place the refractory sheath inside a metal sheath.

Embrittlement due to oil or grease

The couple, the insulators and the inside of the sheath assembly must always be kept clean and free from oil or grease. Most refractories contain some silica, and it has been proved that in a reducing carbonaceous atmosphere the presence of sulphur (frequently present in oil) causes a complex reaction resulting in some reduction of the silica to silicon, which then alloys with the platinum to form a brittle grain boundary constituent. For this reason, thin leather gloves are frequently worn when handling thermocouples to prevent contamination. To obtain good service from platinum thermocouples, cleanliness and adequate protection from contamination must be taken seriously.

Tensile stresses

A platinum thermocouple should also be as free as possible from stress. Table II shows the hardness and tensile strength of platinum and rhodium-platinum alloys used for thermocouple purposes. The breaking load of a pure platinum wire 0·02 in. dia. (the normal thermocouple size) is, therefore, about 7 lb. at room temperature; at 1,400°C. the breaking load is only 12 oz. This represents a load instantly applied. Under prolonged loading, particularly at high temperatures, the wire will withstand only a fraction of this load. It has been shown as the result of creep tests that the safe loading for platinum is about 4·9 tons/sq. in.

Creep tests on thermocouple alloys

In order to compare the resistance to creep at high temperature of platinum alloys normally used for thermocouples, annealed wires, 0·02 in. dia., were suspended under various loads in a furnace at a constant temperature of 1,400°C. The investigation was limited to short-time creep tests. It was shown that, with a load sufficient to fracture the wire in 250 min., a 10% rhodium-platinum alloy will withstand about eight times the loading of pure platinum at 1,400°C., but that above 20% rhodium there is little gain in creep resistance.

TABLE II *Hardness and tensile strength of platinum and rhodium alloys at room temperature*

Rhodium %	Hardness VPN	Tensile strength tons/sq. in.
0	40	10
5	66	16
10	85	21
13	91	23
20	106	27
30	124	30

Heating under load at 1,400°C. produced very large grain growth in the pure platinum, the wire practically consisting of a chain of large crystals of the diameter of the wire. Under prolonged stress at low loads grain boundary sliding occurs and ultimately leads to a brittle type of fracture, which is occasionally found in a couple after long service at a high temperature. Grain growth is far less evident in a 5% rhodium-platinum alloy creep sample, but cracks develop at many of the grain boundaries.

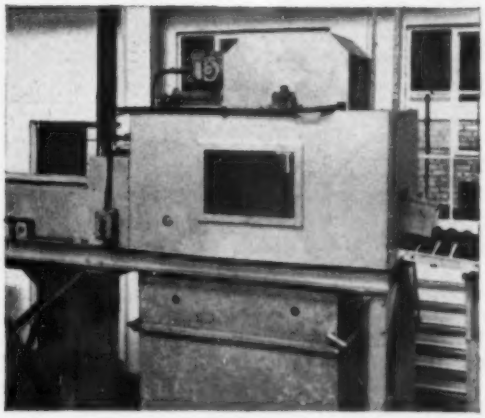
A 5% rhodium-platinum : 20% rhodium-platinum thermocouple, which can be used continuously at temperatures up to 1,700°C., had greatly improved creep resistance by comparison with the platinum : 13% rhodium-platinum couple, the 5% alloy wire having about five times the strength of platinum at 1,400°C. In conditions of extreme loading the 20% rhodium-platinum : 40% rhodium-platinum couple had an even greater advantage, but against this must be set a much smaller e.m.f. of the 20/40 couple and also its increased cost. In practice, however, it is seldom necessary to impose upon the thermocouple anything like these loads.

A thermocouple placed horizontally rests on the sheath, but, if a thermocouple has to be placed vertically, care must be taken to ensure that the weight of the insulators does not apply a tensile stress to the wire. This can be overcome by allowing the hot junction to rest on the sealed end of the sheath. It is undesirable to remove a thermocouple from a hot sheath, particularly if there is any undue friction, as this will impose a considerable stress on the wire. Thermocouples that have failed in service are generally found, by spectrographic examination, to have been contaminated, resulting in embrittlement of the wire. Occasionally, even where there is no obvious contamination, there are clear signs of stress failure.

Spectrographic examination of a 13% rhodium-platinum, said to have failed after two months' service at about 1,100°C., showed no contamination but there was considerable grain growth and sliding at the grain boundaries leading to rupture, typical of slow tensile deformation at high temperature and indicating that the wire had been subject to appreciable stress in service.

Metal degreasing

Double-chain conveyor plant for degreasing by rotation in boiling trichloroethylene



Fully automatic electric-lift mechanism fitted to standard degreasing plant

Since Imperial Chemical Industries Ltd. introduced the trichloroethylene metal degreasing process more than 30 years ago, a speedy, economic and efficient process has been available that can be adapted to suit almost every type of work, provided the contamination is soluble in trichloroethylene or is held to the work by trichloroethylene-soluble materials. The whole cleaning operation is completed in one plant, from which the work emerges in a dry, neutral condition ready for subsequent processes such as heat, chemical or preservation treatment, inspection, repair or further machining.

Methods of treatment vary according to the type of work and the nature of the contamination, and plants can be designed to cater for cleaning by immersion in vapour, immersion in liquor, or combinations of both. In plants designed for vapour treatment the vapour, which is generated by boiling a small quantity of trichloroethylene in the bottom of the plant, fills the plant up to a condensing zone near the top. When a cold metal object is immersed in the vapour, solvent condenses on it and any oil or grease contamination is washed off.

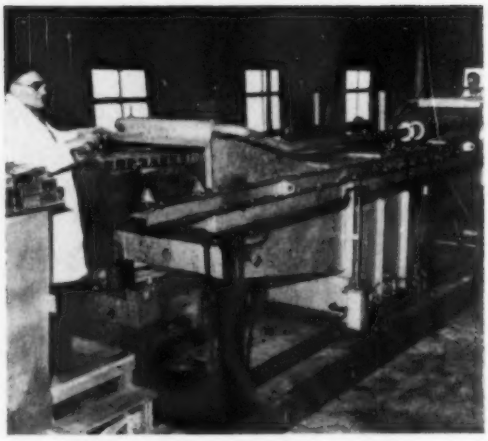
Many types of work are cleaned satisfactorily in this way, but to degrease articles of low heat capacity, to achieve penetration of a bulk of close-lying work, or to effect removal of mixed contamination such as polishing compound, immersion in a bath of liquid trichloroethylene is usually necessary.

The solvent

Colourless, heavy, non-inflammable trichloroethylene, one of the most powerful grease solvents known, is the basis of the process. Its low boiling point (86·7°C.), low specific heat and low latent heat of vaporization ensure only moderate demands on fuel resources and cooling water (*e.g.* the heating required to convert 1 lb. of water to steam will vaporize 8½ lb. of trichloroethylene). The vapour is 4½ times heavier than air.

Although trichloroethylene does not attack the common metals in massive form it may, under certain conditions, react with finely divided aluminium or magnesium. It is liable to decompose, with formation of acid, if exposed to naked flames or other high-temperature sources.

Like most organic solvents, trichloroethylene will cause drowsiness if inhaled in small quantities, and if inhaled in larger quantities, may cause unconsciousness. If handled in a common-sense manner and used in properly designed equipment, however, no trouble should arise.



Continuous vapour plant with mesh belt conveyor

Degreasing methods

Vapour degreasing The simple vapour plant consists essentially of a welded tank of mild steel, galvanized after fabrication, or of stainless steel, fitted with heating equipment and with a condensing coil through which water flows. Solvent is boiled in the sump, filling the plant with vapour to the level of the condensing coil; here it condenses and is collected in a trough, from which it may be diverted outside the plant to a drum or stock tank or allowed to return to the sump. Once weekly, or whenever a quantity of oil has collected in the plant, the solvent is recovered by distillation and may be used again; after cooling, the oily residue is removed through the quick-release cleaning door.

Vapour degreasing is used for straightforward degreasing involving the removal of simple oil or grease contamination.

Liquor degreasing If dirt as well as oil and grease has to be removed, immersion of the work in boiling liquid trichloroethylene is more effective than vapour treatment.

The usual method of providing liquor treatment is to use a plant with either a separate interconnected still or a built-in concentrator; in each case a supply of clean distilled trichloroethylene is led into the bottom of the working compartment, from which contaminated solvent overflows back into the still or concentrator. Most types of small, light articles, which in a vapour plant would reach vapour temperature before degreasing was complete, can be satisfactorily cleaned in a liquor-vapour plant, in which the second compartment can be used for vapour degreasing besides acting as a concentrator.

The plant is similar in construction to the vapour plant but is divided into two compartments, each fitted with heating facilities. In use, one compartment is kept full of boiling solvent and the other is charged in the same way as a vapour plant. Clean distillate returns to the bottom of the liquor compartment, displacing dirty liquor over the weir into the vapour compartment, thus maintaining a self-cleaning action.

Work dipped into the boiling liquor is largely freed from oil, polishing compound, swarf and dirt, and perfect cleanliness can then be obtained by a subsequent immersion in the vapour compartment.

Work handling

To ensure that vapour degreasing is effective, and to avoid dragout of solvent either as liquid or vapour, it is essential to load all work in such a position that it will drain freely. Some articles, such as heavy machinery parts, are conveniently degreased while directly suspended from a hoist. Most other types of work are best handled in baskets, which can rest on the workplates in the plant. After reaching vapour temperature, work is removed from the plant slowly and steadily.

Small articles such as castings with blind holes or cup-shaped pressings, which will trap solvent, are loaded into cylindrical baskets that can be turned over before they are removed from the vapour.

Metal drying

It is well known that wet metal articles can be dried by immersing them in the vapour of a boiling chlorinated solvent, such as trichloroethylene—the solvent used in most I.C.I. degreasing plants. The moisture film dries off because, although the boiling point of trichloroethylene (87°C.) is below that of water, a mixture of the two boils at an even lower temperature (73°C.). Vapour degreasing plants can therefore be used for drying small metal articles, although the evaporation of large drops of water from the surfaces of the articles usually leaves stains.

A new I.C.I. technique not only dries more rapidly than previous methods, but without the formation of stains. Trichloroethylene is still the drying medium but the treatment is in liquor instead of vapour, and to prevent stains forming on the articles a special metal drying additive—'Trisec'—is mixed with the solvent. Only a small proportion needs to be added, about one part to 80 parts by volume of trichloroethylene. In action, a film of trichloroethylene containing 'Trisec' displaces the water film on the metal surface, so that the metal emerges from the final rinse with a film of pure solvent which, during removal from the plant, evaporates leaving no stains.

PEOPLE

THE ROYAL SOCIETY has been informed that **Dr. Louis Essen,** a senior principal scientific officer at the National Physical Laboratory, has been awarded the A. S. Popov's Gold Medal by the Academy of Sciences of the USSR. This award is for the most distinguished scientific work in the field of radio-engineering performed during the period from 1956 to 1958. The work of scientists from all countries is considered, but this is the first time that it has been awarded to a scientist outside the Soviet Union.

The outstanding achievement which has won Dr. Essen this recognition has been his work leading to the establishment of an atomic frequency standard as a possible basis for the future standard of time which would have greater precision and be more easily accessible than the astronomical standard for the purposes of radio-engineering and physics.

Dr. Essen was educated at the University of Nottingham, and joined the scientific staff of the National Physical Laboratory in 1929. In his early years at the Laboratory he developed the Essen quartz-ring-oscillator which is widely used in frequency-standardizing laboratories and observatories. From 1940 to 1946 he worked on radio measurements at micro-wave frequencies and since then has applied the newly-developed techniques to the field of precise measurements and standards. In 1953, with J. V. L. Parry, he started work at the National Physical Laboratory on an atomic standard of frequency.

Dr. W. Steven, superintendent of the Development and Research Department Laboratory of the Mond Nickel Co. Ltd., Birmingham, has been transferred to the Development and Research Division of the International Nickel Co. Inc., New York, as director of research. He has also been elected an assistant vice-president of that company.

Dr. Steven took a 1st Class Honours B.Sc. degree at Glasgow University in 1939 and obtained his Ph.D. degree in 1942. After five years with William Jessop & Sons Ltd., Sheffield, he joined the Mond Nickel Company in 1947 as a research metallurgist. In 1955 he was appointed superintendent of the Laboratory.

Dr. Steven is a Fellow of the Institution of Metallurgists, and is a member of the Iron and Steel Institute, the Institute of British Foundrymen and other societies.

Dr. G. L. J. Bailey has been appointed superintendent of the Development and Research Department Laboratory of the Mond Nickel Co. Ltd. in Birmingham.

Dr. Bailey took a B.Sc. (Hons.) degree at Imperial College in 1937 and obtained his Ph.D. in 1939. After experience at the Admiralty Engineering Laboratory, West Drayton, he joined the British Non-Ferrous Metals Research Association, where he became deputy research manager. Since 1955 he has been superintendent of the Platinum Metals Research Laboratory of the Development and Research Department, the Mond Nickel Co. Ltd., Acton.

Dr. Bailey is a Fellow of the Institute of Physics and a Fellow of the Institution of Metallurgists. He is a member of the Institute of Metals, the Institute of Metal Finishing and other societies.

Dr. W. I. Pumphrey, D.SC., has been elected to the Court of Governors of the University of Birmingham as a representative of the Guild of Graduates. Dr. Pumphrey is manager of the Research Department of Murex Welding Processes Ltd.

Dr. R. T. Parker

Dr. Robert T. Parker has been appointed head of Aluminium Laboratories Ltd., Banbury and Geneva offices, in succession to Mr. R. D. Hamer.

Dr. Parker, an Associate of the Royal School of Mines, joined the Aluminium Ltd. Group in 1938 and transferred to Aluminium Laboratories Ltd. in 1946 as head of the Metallurgical Division. In 1950 he became Director of Research at the Banbury laboratories. He had previously been engaged in research with the British Non-Ferrous Metals Research Association. Dr. Parker is a Fellow of the Institution of Metallurgists, a Fellow of the Royal Institute of Chemistry, and chairman of several committees in the aluminium industry.

Dr. W. Betteridge has been appointed superintendent of the Platinum Metals Research Laboratory of the Development and Research Department, the Mond Nickel Co. Ltd., Acton.

Dr. Betteridge was educated at King Edward's School, Birmingham, and took a 1st Class Honours degree in Physics at Birmingham University in 1931, obtained his Ph.D. in 1933 and his D.Sc. in 1958. After serving with Joseph Lucas Ltd. and the Bristol Aeroplane Company he joined the Mond Nickel Company in 1947, where he has been concerned primarily with the development of high-temperature materials, at first in the Birmingham Research Laboratory and more recently as development officer in London.

Dr. Betteridge is a Fellow of the Institute of Physics and has presented numerous papers on high-temperature materials and other subjects.

Two new appointments have been made in the Meter, Relay and Instrument Division of the English Electric Company. **Mr. G. E. Robertson,** A.M.I.E.E., ASSOC. MEM.A.I.E.E., has been appointed manager, Sales and Contracts, of the Meter, Relay and Instrument Division, and **Mr. P. G. Bevis** has been appointed manager, Sales and Contracts, Relay Department.

Mr. J. G. Wistreich, M.SC.(ENG.), D.I.C., A.M.I.MECH.E., head of laboratories, BISRA, Sheffield, has just received the degree of D.SC.(ENG.) from the University of London for his work in metallurgical engineering, with particular reference to wire drawing.

The Council of Industrial Design announces the appointment of **Mr. J. Noel White** to be deputy director as from January 1, 1960, when Mr. Paul Reilly, the present deputy, succeeds Sir Gordon Russell as director of the Council.

The Council of Industrial Design has appointed **Mr. W. H. Mayall** as industrial officer for the engineering industries. He takes the place of Mr. L. A. Grosbard, who has now joined Mullard Ltd.

Mr. Mayall's main responsibilities will be in the capital goods field. At the recent Engineering Design Conference in Birmingham, ways were suggested in which the engineering industries and industrial designers could co-operate more closely. Mr. Mayall will use these suggestions and the other recommendations of the Conference as the basis of his work.

Before joining the COID Mr. Mayall was project engineer and later mechanical engineer to Tiltman Langley Ltd. He has been concerned with mechanical and aeronautical work since he joined Flight Refuelling Ltd. in 1942.

Mr. J. F. Robertson, C.A., a director and group treasurer of the Hawker Siddeley Group, has been appointed to the board of High Duty Alloys Ltd., a member of Hawker Siddeley Industries Ltd. Mr. Robertson joined the Group in 1938 and was appointed to the board in 1953.

Another new appointment to the board of High Duty Alloys is that of **Mr. Ian C. Dick,** C.A., who joined the company in June, 1952, as chief accountant and was appointed company secretary in February, 1954, a position which he still retains.

Mr. I. H. Gordon has been appointed assistant sales manager of Cambridge Instrument Co. Ltd. at the company's head office in London.

Mr. Gordon was previously with Land Pyrometers Ltd., and prior to that was with Evershed & Vignoles Ltd.

Mr. C. A. B. Malden has been appointed joint managing director, with Mr. J. G. Cronk, of Amber Oils Ltd. He joined the company in 1958 from Stephenson Clarke Ltd. Mr. Malden is a Fellow of the Institute of Directors and a director of Finders Services Ltd.

Mr. Victor Brenner has been re-elected president of the Birmingham Engineering Centre for the fourth year and **Mr. C. J. Grazebrook** and **Mr. A. G. Banks** have been elected vice-president and treasurer respectively.

Mr. Leslie J. Cox, M.I.P.R., has been appointed chief of the P.R. Department of Wolf Electric Tools Ltd. The functions of this department will also cover press relations, market research (home), employee relations, sales training and education.

Mr. Edward Patterson, M.I.B.E., M.S.I.A., has been appointed to the newly-created post of technical liaison officer of Wolf Electric Tools Ltd. In addition to internal technical committee work he will represent the company on all British and International technical committees. Mr. Patterson will also be responsible for the production of instruction folders and the handling of technical queries, new ideas, etc.

Longer-term grants for special researches

Longer-term grants for special researches will be awarded by the Department of Scientific and Industrial Research in future. These grants—which have normally been limited to a maximum of five years—may now be awarded for an initial period of up to seven years with the possibility of continuing support for a further 10 years.

An experimental plant for the production of gallium has begun its test run at Ajka, eastern Hungary. The gallium (worth about £35 an ounce) is salvaged from the caustic solution used in the extraction of alumina for the aluminium industry.

LETTERS *to the Editor*

Germanium

SIR: The article which appeared in your March issue under the title 'Germanium' contained more absurd statements than can be allowed to pass unchallenged. I presume that the article was based on a translation of the original German, but there is no excuse for employing such atrocious broken English that the reader is unable to obtain an understanding of the processes described.

I defy anyone to interpret the following extract unless he already knows the process: '. . . Any impurities which may still be present are eliminated in the so-called "zone melting process," a crystallizing process in which the glow wire of a high-frequency transmitter is passed through a germanium crystal lying horizontally. The impurities in the fusion zone of the crystal are concentrated by this means and, since with the new structure of the crystal lattice all foreign matter is segregated, by frequent repetition of the process, very pure germanium is obtained.'

Bad translating, however, is not responsible for the objectionable bias of the article. The ludicrous distortion of the history of the development of germanium devices is even carried to the point of insinuating that the Americans carried off the secrets of germanium rectifiers—if not of the 'transistor' itself—as spoils of war. In fact, of course, the brilliant American research work which culminated in the post-war discovery of transistor action was well advanced by the year 1944, and the high back-voltage germanium rectifier was being manufactured (H. C. Theurer and J. H. Scaff, N.D.R.C. interim report No. 1, Contract O.E.M.S.R.—1408, November 21, 1944) by processes which employed many of the techniques that are in use at the present time.

S. E. BRADSHAW

Research Laboratories of the General Electric Co. Ltd., Wembley, Middlesex.

NEW FILMS

Stelvetite

The story of Stelvetite, the new plastic-coated steel sheet produced by John Summers & Sons Ltd., of Shotton, Chester, has been chosen by the Rank Organization for one of the first of a new weekly series of magazine films in colour.

The new series, called 'Look at Life,' has just been introduced to replace the black and white newsreels of the Rank Organization shown at Odeon, Gaumont and many independent cinemas.

The development story of Summers' Stelvetite occupies the whole of one of the 10 minute reels. Some outstanding colour shots filmed in a large steel works are contained in the film. It shows the manufacture of the PVC plastic at the Manningtree factory of B.X. Plastics Ltd., and then describes the making of the steel sheet at Summers' works at Shotton, Chester, and the subsequent production of the Stelvetite.

The Rank Organization filmed, for the first time, the Stelvetite production line where, as the title of the film suggests, 'A Marriage is Arranged' between the steel sheet and PVC plastic.

The story finishes with scenes showing how this new material is already being used for a wide variety of domestic articles and by telling of its future possibilities.

For Temperature Measurement and Control

the KH INDICATING CONTROLLER MARK 4

This advanced Indicating Controller incorporates extremely accurate measuring and control sections in a single compact case. It utilises a photo transistor and a moving coil micro-ammeter thus dispensing with the need for thermionic valves. Requiring less than 46 sq. in. of panel space, the unit has a calibrated indicating scale 8 in. long.

A complete range of instruments is available, which meets requirements over a wide field of process control applications, including models with 2-Position, Proportional and 3-Position action. A programme feature can be incorporated.

- **No Thermionic Valves**
- **Automatic Safety features for Plant Protection**
- **Control Sensitivity: better than 0.1% of F.S.D.**
- **Calibrated Accuracy: within 1% of F.S.D.**
- **Overall accuracy in accordance with B.S. 1041/1943**
- **Plug-in unit with automatic location**
- **Panel space required: only 10½ in. x 4¼ in.**
- **Competitively priced**

Please write for Kelvin Hughes publications IND. 703, IND. 704 and IND. 705 which cover 2-Position, Proportional and 3-Position instruments respectively and are available on request.

KELVIN HUGHES PYROMETRY INSTRUMENTS
include
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FOUNDRY EXHIBITION

SIR FREDERICK SCOPES opened the recent Foundry Exhibition at Bingley Hall, Birmingham. The exhibition, the first of its kind devoted solely to the foundry industry, attracted nearly 14,000 visitors and nearly 100 firms took part in the display.

Exhibitors

The Incandescent Group focused attention on one piece of equipment, relying on models and photographs to illustrate other designs of foundry plant made by the group.

The main feature was a full-size independently-fired air heater for producing hot blast. This heater is designed to supply blast for two 4-ton-per-hour cupolas. The working temperature (500°C.) is reached within 30 minutes of starting. Heaters already installed have shown savings: in one case the cost of the installation was recovered in less than 10 months' operation. Details of construction were shown by a scale model of the heater with cutaway sections, showing the oil burner and heat exchanger tube bundle.

Also on the stand were two models: a direct-arc melting furnace and a twin cupola installation. The Metalectric arc furnace incorporates hydraulic control for all motions and is built with capacities from 500 lb. to 150 tons. The twin-cupola model reproduces a plant which has automatic charge make-up and recording. The cupolas are fed by an automatic swivel charger. The molten iron is tapped into a 4-ton oil-fired hot metal receiver. Incandescent-Whiting wet-dust arresters are fitted to both cupolas.

Also on display was a model of bogie-hearth furnace with Laclede suspended roof, suitable for heat treatment of castings.

The newly-formed company of **Birlec-Efco (Melting) Ltd.** was showing two working furnaces. One was a self-contained high frequency induction melting unit. The other was a holding furnace for the automatic pouring of molten aluminium and zinc-base alloys.

Power for the high-frequency melting furnace is provided by a 15 kW. 'power pack' which can be used for melting and induction-heating operations such as hardening, annealing, sintering, brazing, pre-heating, hot-forming, and vacuum heating processes. At this exhibition the unit was used in conjunction with a melting furnace and small castings in steel and iron were made.

A Birlec Ajaxomatic furnace was being demonstrated in use with E.M.B. diecasting machinery. A holding unit for pouring molten aluminium and zinc-base alloys, automatic operation enables the rate of diecasting production to be raised considerably. By eliminating ladling it ensures far greater metallurgical control.

Similar in principle to the Tama Mains-Frequency Channel-Type Induction-Melting furnace it is rated at 20 kW. with a molten capacity of 300 lb. and an average power consumption of 10-12 kW.

Also on show was a working model of a Detroit Rocking-Arc furnace. This type of furnace is used primarily as a batch melting unit for a wide variety of applications in ferrous and non-ferrous foundries. The melting chamber is rotated backwards and forwards during melting and the charge is heated by radiation from the arc between the electrode tips and by conduction from the heated refractories.

Two crucible furnaces were shown on the stand of

Boys from the National Foundry Craft Training Centre, West Bromwich, were responsible for one of the displays at the Foundry Exhibition.
A score of students, average age 16/17, working in shifts in teams of four, manned a complete foundry in miniature and demonstrated the type of work and training carried out at the Centre

The Morgan Crucible Co. Ltd. They were the Birlec Morgan Electric Diecasting Furnace Mark II and the oil fired Morgan Basin Tilter.

The Birlec Morgan E.D.F., developed only a year ago, was shown in an improved Mark II version. The furnace can handle a throughput at 720°C. of approximately 125 lb. of aluminium per hour.

The oil-fired Basin Tilter as a bulk melter gives simple, comfortable operation, all the advantages of lip pouring and a high melting rate. This furnace is the most up-to-date unit so far produced for feeding maintaining furnaces in the die foundry. It has a 380 lb. or 500 lb. aluminium capacity, depending on the crucible used, and can bring down a 380 lb. charge in 30 minutes or 500 lb. in 40 minutes. It can also melt $\frac{1}{2}$ ton of brass in an hour or $\frac{3}{4}$ ton in 80 minutes. Similar charges of gunmetal take only 10 minutes more in each case.

The Mark II miniature furnace, now supplied with an industrial-type fan, was also on display together with a range of Suprex Crucibles.

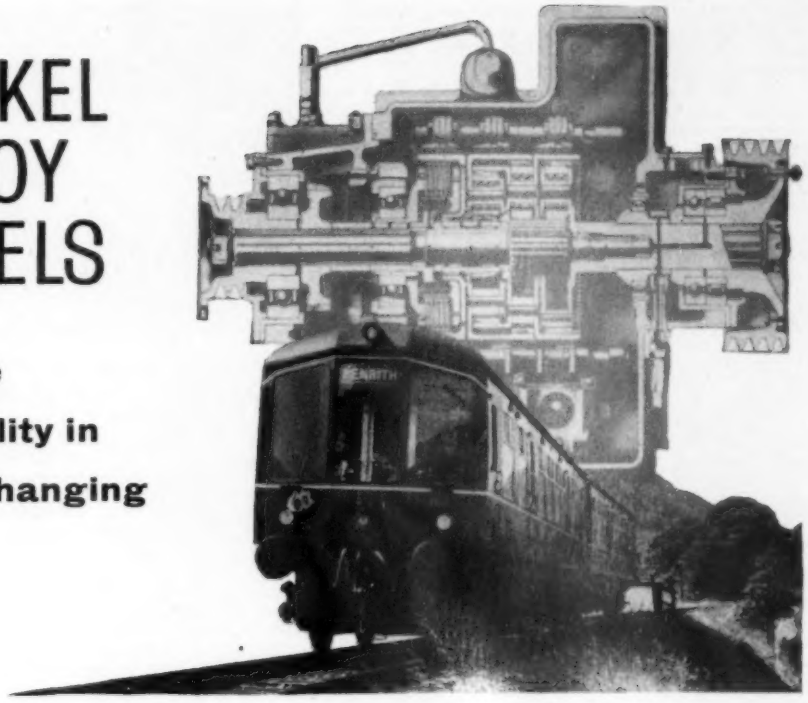
Four I.C.I. Divisions—Billingham, Alkali, Heavy Organic Chemicals and Metals—and Marston Excelsior Ltd., an I.C.I. subsidiary, were on show.

Metals Division, Marston Excelsior Ltd. The exhibits covered by this firm were wrought and fabricated titanium products for nuclear engineering, heat transfer materials and specialized products.

Of particular interest in the titanium section were exhibits featuring platinized titanium anodes (including

NICKEL ALLOY STEELS

ensure reliability in gear changing



The self-changing gear box, produced by Self-Changing Gear Box Ltd., Coventry, is standard on road, rail and water transport in many lands, and notably on London's buses.

This famous unit, in both its pre-selector and fully automatic versions, incorporates a range of nickel alloy steels. Although there is no direct mechanical operation of the gears in the normal way, heavy train, truck and bus operation inevitably imposes severe stresses upon the gears and to withstand these stresses most of the operating mechanism in the 'Wilson' box is made from nickel alloy steels, including the 3 per cent nickel-chromium-molybdenum EN 36 case-hardening steel.

TYPICAL CORE MECHANICAL PROPERTIES OF

EN36

SIZE	HEAT TREATMENT	MAXIMUM STRESS t.s.i.	ELONGATION per cent	IZOD ft. lb.
1½" dia.	Oil quenched 780°C.	72·0	19	69
2½" dia.	Oil quenched 860°C. Oil quenched 780°C.	62·7	19	67
3" dia.	Oil quenched 860°C. Oil quenched 780°C.	59·7	22	72

The benefits to be gained from the more highly alloyed case-hardening nickel steels, such as EN 33, EN 34, EN 36 and EN 39 include ease of heat-treatment, minimisation of processing distortion, and general reliability

Send for 'The Mechanical Properties of Nickel Alloy Steels'

MOND NICKEL

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YOA 21/58

copper-cored rod) for the cathodic protection of vessels, pipelines and jetties. Titanium turbine blades, power station condenser tubes and funnel exhaust scrubbing equipment were also seen.

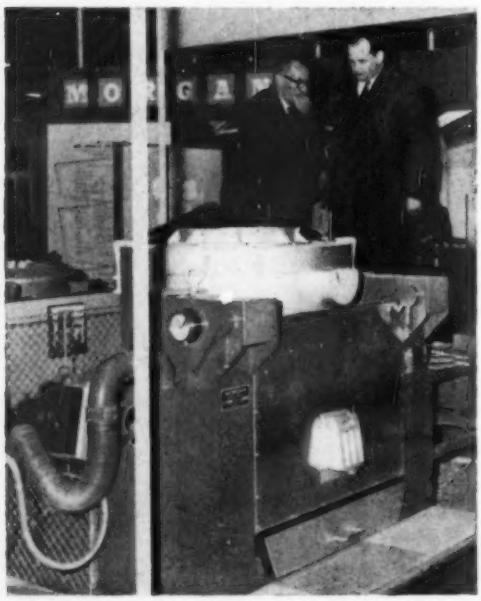
The ' new ' wrought metals such as zirconium, titanium, niobium and vanadium were shown in the nuclear engineering section. This display also includes I.C.I. fuel cans and ' Integron ' extended surface boiler tubes.

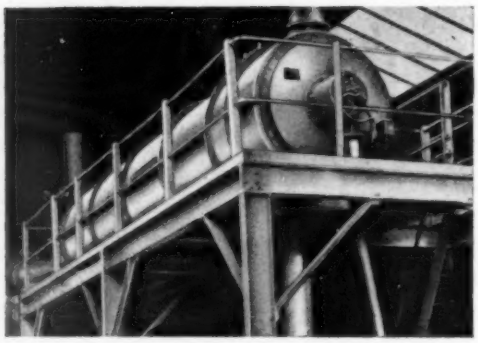
In the section devoted to heat-transfer products, typical aluminium alloy and titanium heat exchangers fabricated by Marston Excelsior Ltd. were shown.

The Alloys Division of Union Carbide Ltd. showed many of the ferro alloys and metals for supply to the iron, steel and non-ferrous foundries, together with a special display of technical data on rarer metals and their compounds. Emphasis was given to exothermic alloys, calcium manganese silicon, and silico manganese. The production of steels using exothermic alloys as ladle additions is now common practice and details of the latest alloys and techniques for using them were available.

CIBA (A.R.L.) Ltd. was showing a new method for vacuum forming large foundry patterns of complex shapes by a ' wet lay-up ' method using ' Araldite ' epoxy resin and glasscloth. The vacuum table used has an airtight bottom forming a box into which the vaccum pipe is led. Slotted vents are placed at intervals in the top plate. A negative of the object to be made is placed on the table and, after applying a release agent, an ' Araldite ' gel-coat is applied. A piece of glass-fibre mat is saturated with unfilled ' Araldite ' resin and placed over the gel-coat. Three or four layers of dry glass-fibre mat are placed on top of this. The whole is then covered with a sheet of polyvinyl alcohol, which is secured round the edges.

Below is seen a tilting furnace shown on the stand of the Morgan Crucible Co. Ltd.





The direct-fired cupola blast heater featured on the stand of the Incandescent Group

When a vacuum is applied, the P.V.A. sheet is drawn down and follows the contour of the negative. The excess resin in the first layer of glass-fibre mat is forced up through the dry layers. The vacuum is maintained until the resin has set. The result is an air-free compact laminate of ratio about 50:50 glass to resin. The shell formed by this method is backed with a filled ' Araldite ' casting resin.

The Shell Chemical Company stand illustrated the use of ' Epikote ' epoxy resins in the manufacture of loose patterns, pattern plates and core boxes.

Lafarge Aluminous Cement Co. Ltd., manufacturers of Ciment Fondu Aluminous Cement and Secar 250 White Calcium-Aluminate Cement, showed some of the uses for refractory concrete made with these materials in the foundry and allied trades, including floors, annealing furnaces, melting furnaces, core stoves, cupolas, etc.

Furnaces and combustion chambers lined with refractory concrete made with these cements were on view as well as many complex shapes.

In demonstrations of the high thermal shock resistance of refractory concrete, specimens heated to a white heat were plunged halfway, and then fully, into cold water to demonstrate convincingly the non-spalling properties.

Mellor Mineral Mills were exhibiting a wide range of Dohm foundry products. They included an odourless ignitable mould wash called DOHMspray; DOHMseal refractory glazes; DOHMpatch plastic refractory; graphitic linings and the new Investment Base 350 for precision investment casting which appeared on the market recently.

Investment Base 350 was developed after tests on the American cristobalite base and operates at a temperature of 1,450°C. It needs no additives, retarders or accelerators, only water to mix into a slurry. Setting time is 20 minutes.

Foundry Services Ltd. showed the making of production die-castings and, on the non-ferrous side, the making of identical castings, one with a FEEDEX exothermic feeder-head and one conventionally fed; one treated with a mould coating and one without any dressing.

Various other FOSECO products were demonstrated, among them the POROTEC apparatus for the rapid detection of gas in melts of aluminium alloys, the portable hand-mixer, spray guns and the new TAK extruder—TAK being a new compound for sealing moulds to prevent run-out.

Refractory Concrete

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Material
made with
crushed firebrick
and
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best for Bogie Tops

because: The reduction of joints eliminates displacement troubles.
It is ready for use and of great strength and hardness within 24 hours.
—It will withstand severe thermal shock without spalling.
It has a smooth level surface. It has no appreciable drying shrinkage or after-contraction.
It is stable under load up to 1350°C. (2460°F).

Write for further details and photographic examples.

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USE SECAR 250
(an iron-free white calcium-aluminate cement)
for
Super Duty and Special Conditions of:
Higher temperatures up to 1800°C
Reducing atmospheres
Resistance to Slag attack
Resistance to products of combustion
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73 BROOK STREET, LONDON W.1. Telephone: MAYfair 8546

NEW PLANT

Humidity cabinet equipped for refrigeration

The Hedin standard range of humidity cabinets, made by Hedin Ltd., Commerce Estate, South Woodford, London, E.18, now includes a model equipped with a refrigeration plant to enable low-temperature testing to be carried out to M.O.S. and other specifications.

The unit has a capacity allowing temperatures of −10°C. to be obtained in an ambient of 20°C. Automatic temperature control is provided by an adjustable thermostat. The unit is powered by an electrically-driven, hermitically-sealed compressor circulating Freon 12 refrigerant. Refrigeration can also be arranged to provide dehumidification at ambient temperature.

Construction is of stainless steel and aluminium with electro-plated brass fittings to eliminate the possibility of rusting. Fibreglass 2-in. thick ensures high thermal stability and efficiency. A practically silent fan unit provides an ample air flow through the humidity chamber, giving a maximum temperature gradient of ±1°C. throughout the working space.

Humidity is normally produced by steam injected into the air stream near the fan and thoroughly mixed and cooled before it reaches the humidity chamber. If extended ranges of humidity or temperature are required, a comprehensive set of auxiliary equipment is available, in addition to refrigeration, including water cooling and water spray. Timers can be fitted providing automatic control at predetermined conditions such as are required by Ministry of Supply specifications.

Cold-bend testing machine

A motor-driven cold-bend testing machine, suitable for bending iron or steel bars up to 2 in. square or plates up to 6 in. wide, has recently been made in this country. The machine was built under licence from the Tinius Olsen Testing Machine Company, Philadelphia, U.S.A., by Edward G. Herbert Ltd., Manchester, 19, who act as sole agents in Great Britain and Eire.

The specimen to be tested is bent to a maximum angle of 180 degrees round a centre pin, five standard sizes of pins being used. The pusher pin, which is mounted on a substantial table, is readily adjusted in relation to the centre pin so that a perfectly round bend may be obtained. A graduated scale is provided round the moving table so that the angle of bend of the specimen may be noted at any time during the test.

2 *Cold-bend test machine*

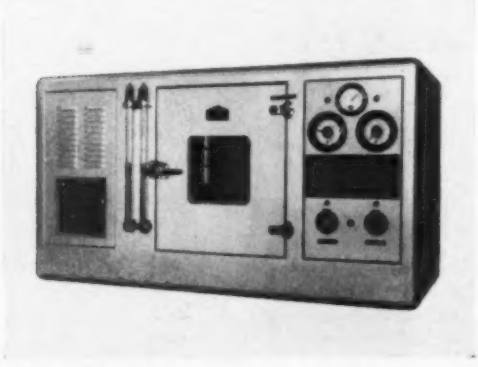
The load is applied by a 5 h.p. motor through a worm gear reduction to the table which is a steel casting. Heavy steel plate and wire-mesh guards are provided to protect the operator against possible injury from fracture of the specimen during the test.

Lapping attachment for milling cutters

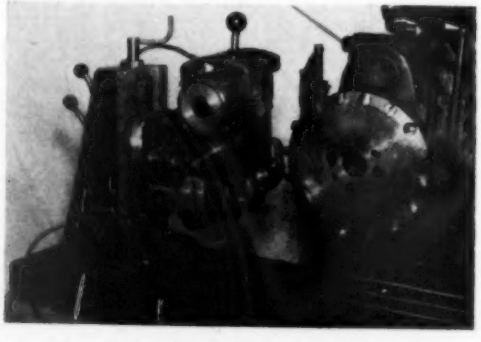
To ensure flatness and fine finish in face milling, it is essential that the cutter teeth operate in a true plane parallel to the work surface and that their cutting edges are very finely lapped. When these factors are coupled with modern milling machines, resultant surface finish is such that subsequent grinding operations are often unnecessary.

To obtain the necessary relationship between the cutter and the workpiece it is necessary to lap the cutter in position on the spindle from the machine table. This ensures that any inaccuracy in the spindle and cutter is corrected relative to the movement of the machine table surface and the face of the cutter.

To achieve this, an attachment which bolts directly to the machine table has been designed by Alfred Herbert Ltd. This device, the Herbert Milling Cutter Lapping

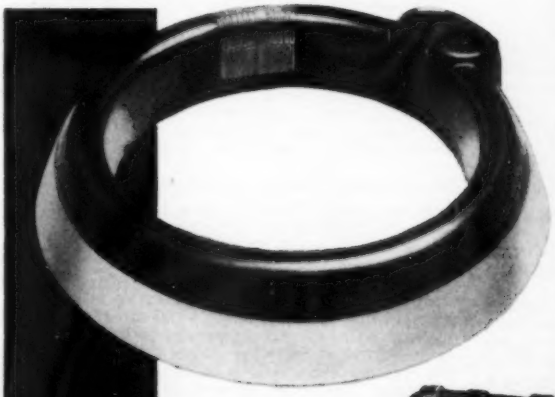


1 *Hedin humidity cabinet*



3 *The A. Herbert milling cutter lapping attachment*

HERE'S RELIABILITY FOR YOU



The reliability of Brightray materials is proved in exacting situations like these. The Brightray series includes suitable resistance alloys for most industrial and domestic purposes, including electric fires, immersion heaters, boiling rings and other appliances.

DE-ICING AIRCRAFT
The Rolls-Royce Dart engine intake in the Vickers Viscount is equipped with a Dunlop electrothermal de-icing system incorporating elements made from BRIGHTRAY S and FERRY.

SALT BATH HEATING
This immersion heater, for the exacting task of industrial salt heating, uses BRIGHTRAY C electrical resistance alloy. (British Thomson-Houston Company Limited)

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NICKEL-CHROMIUM Electrical Resistance Alloys

Table:

BRIGHTRAY C for intermittent heating up to 1150°C . . . used in cookers, toasters, laundry irons, fires, soldering irons . . .

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BRIGHTRAY S for continuous heating up to 1150°C . . . mainly for electric furnace elements

BRIGHTRAY F for electric furnace elements operating in special atmospheres up to 1000°C

BRIGHTRAY H for furnace elements in the range 1100 to 1250°C (produced in strip form only)

FERRY for resistances on electrical control gear and instruments up to 300°C

WILCO-WIGGIN THERMOMETALS in many grades for thermostatic devices controlling temperatures ranging from —75 to 540°C

Produced in the form of Rod, Wire, Strip and Tape

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Attachment, enables the operator to obtain the true relationship, and also to restore, quickly and easily, the fine lapped finish to the blades. The lapping operation takes place on the machine without the setting being altered, with a minimum of idle time. The attachment is also portable, enabling it to be readily moved from one machine to another.

The attachment is available as a single wheel unit. Provision is made on the attachment for fitting the lapping wheel at either end of the spindle, enabling lapping to be carried out in any position within the limits of the attachment when used on vertical or horizontal single-spindle milling machines. A separate wheel and indexing box can be provided, converting the attachment into a more universal one for use on Duplex or multi-spindle machines, where it is not always convenient to change the position of the wheel

Radiant flash annealing furnace

At its Dolgarrog Works in North Wales the Aluminium Corporation Ltd. is operating a new type continuous flash annealing furnace in which circles of aluminium manganese alloy are heated rapidly by radiation, to produce the fine grain structure needed for deep drawing. Results comparable with salt-bath treatment are being achieved with freedom from distortion, staining and surface damage. The radiant flash annealing furnace was designed and manufactured by Royce Electric Furnaces Ltd., Walton-on-Thames, Surrey, in conjunction with the Aluminium Corporation. The radiation is obtained from heating panels which are fitted with heavy gauge nickel-chrome heating elements and positioned in the roof and the hearth of the furnace.

Variations in heating rates are readily made by adjusting the spacing of the panels from the work, the roof and hearth being built as separate units suspended from the main framework of the furnace and contrabalanced. These units are moved by synchronously-driven jacks operated by handwheels which allow fine adjustments to be made. The furnace can be opened in this way to give easy access to the interior and the heating elements. A gauge is provided to show the spacing of the heating panels from the charge. Vertical sliding doors at both ends of the furnace allow the entrance and exit to the heating chamber to be closed as far as possible to reduce heat losses.

The charge is carried through the furnace on a horizontal conveyor which has two endless chains running along the sides of the chamber and carrying cross-flights of heat-resisting steel. The cross-flights are covered with woven asbestos cloth to provide a non-abrasive support. The conveyor is driven by a squirrel-cage motor through reduction gear, its speed being variable between 4 and 30 ft./min. In the design and control of the furnace special precautions are taken to avoid temperature fluctuations. The elements are supported in grooved refractories of high density which, having a high heat content, serve as a heat reservoir. The temperature of the furnace is controlled in four independent zones, the automatic indicating controllers being used not to switch the element supply but to regulate the supply voltage. The conveyor temperature is maintained as far as possible when it leaves the heating chamber by providing a closed and heat-insulated return path.

Safety precautions are incorporated in the furnace to protect the elements from damage by overheating, and to protect a charge should the conveyor be stopped accidentally. In this event the supply to the elements is removed automatically and an aural alarm is given. The furnace has a rating of 90 kW., it provides temperatures up to 900°C., and has a heating chamber 21 in. wide by 15 ft. long. Heating times for circles ranging from 6 in. to 20 in. dia. and from 0·02 in. to 0·08 in. thick vary from 1½ to 4 min. The output of circles of 18-gauge material, 12 in. dia., is a little over 2 cwt./h.

New tool steel

The Tool Steel Department of English Steel Rolling Mills Corporation Ltd., Openshaw, Manchester, is producing in increasing quantities a graphitic-type tool steel. Known as 'M.I.C.8,' it is an oil-hardened tool steel having freedom from distortion in heat treatment equal to that of conventional carbon-manganese 'non-distorting' steels. It differs, however, from these conventional steels in that it contains a higher carbon content, and by careful control of the analysis and metallurgical conditions of processing a portion of the carbon is retained in the final product in the form of graphite particles uniformly dispersed throughout the structure.

The graphite particles act as a lubricant and considerably improve its machinability. They also reduce the tendency to seize, gall, or score induced by the rubbing of metal against the surface of the tool. These properties combined with its uniform response to heat treatment and good hardenability make it ideally suited for components of intricate shape which have to withstand wear and abrasion. It is produced as rolled or forged bar, forged blanks and rings.

New gas torch for high-temperature welding

A gas/air torch which provides a small diameter, highly concentrated flame, ideal for precision aluminium welding, silver soldering, copper brazing, etc., has been introduced by Sunvic Controls Ltd. The torch uses coal gas at normal mains pressure; apart from convenience this, it is claimed, reduces operating costs to approximately one-third those of an oxy-acetylene torch.

A preheater raises temperature of both gas and primary air to 750°C. before they reach the mixing chamber. The resultant flame cannot be blown out and continues to operate even when the torch is played into an inert atmosphere. The inner section of this flame, approximately 4-5 cm. long, has a temperature greater than 2,000°C. It will readily melt 1 mm. platinum wire.

When the torch is used for silver soldering stainless steel there is a noticeable reduction in distortion. It has been successfully applied to soldering and brazing high magnesium alloys and also for autogenous welding of brass aluminium and A1-Mg alloys.

Automatic pre-treatment plant

Of very flexible design and adaptable for various types of cleaning and process treatments, including pickling and phosphating, the 'Agi-Dip' fully automatic multi-stage pre-treatment plant has recently been introduced by the Electro-Chemical Engineering Co. Ltd., of Woking, Surrey. In this type of plant the parts to be treated are loaded into baskets which are then automatically passed through agitated immersion stages, the number of these depending on the treatment required. Trichlorethylene vapour degreasing, spray rinsing and drying sections can be incorporated into the sequence if required. Agitation is carried out by means of a reciprocating pneumatic arrangement adjustable for amplitude and frequency of stroke. Double-acting air cylinders enable the work platforms to be driven down into the solutions at a faster rate than the freely falling components, thus effecting complete separation of tightly packed parts. The work is transported through the sequence by means of an index beam conveyor controlled by a timing mechanism. This mechanism can also operate automatic loading and unloading platforms to link the plant with a shop conveyor. Only one operator is normally required for the plant.

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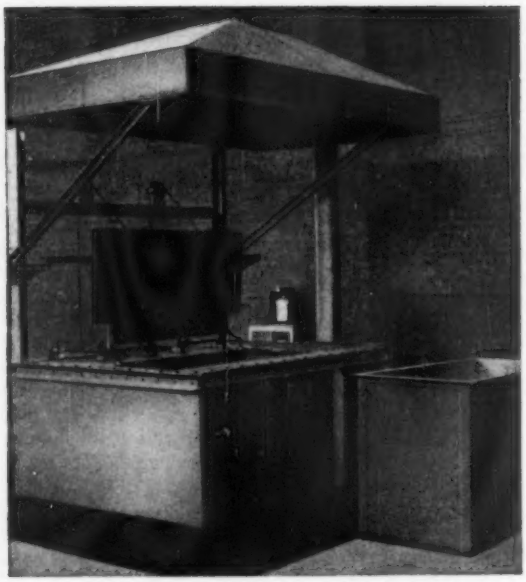

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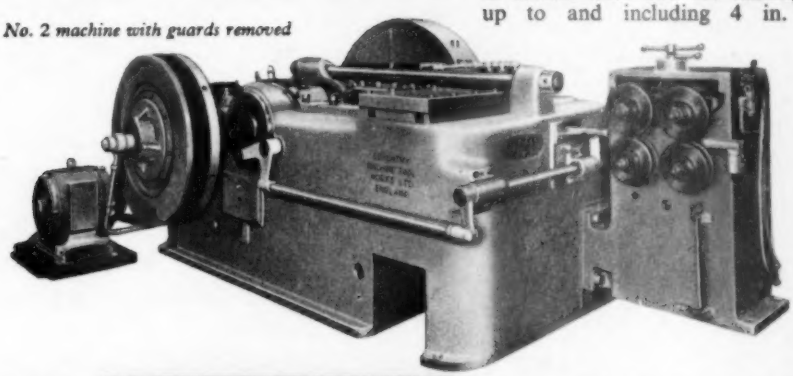

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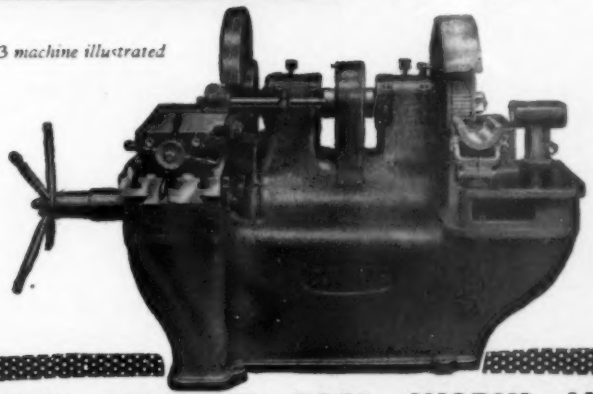

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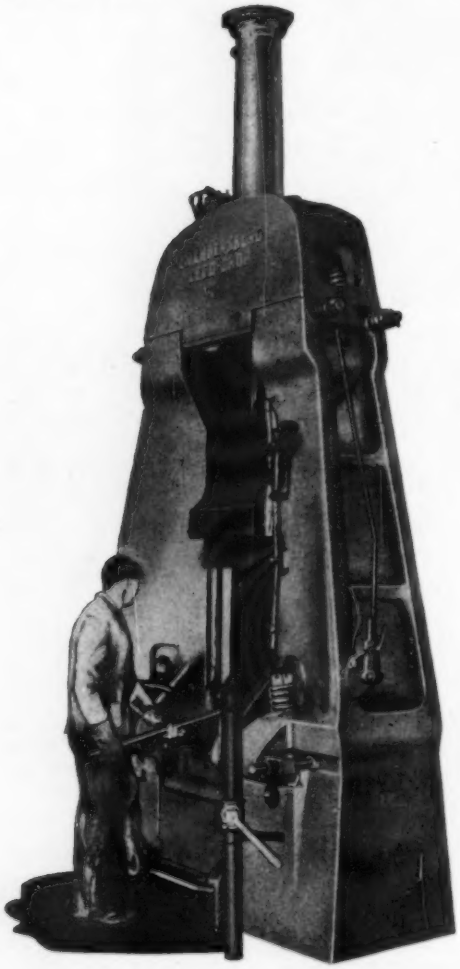

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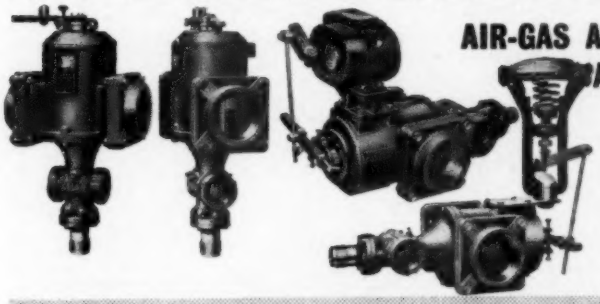

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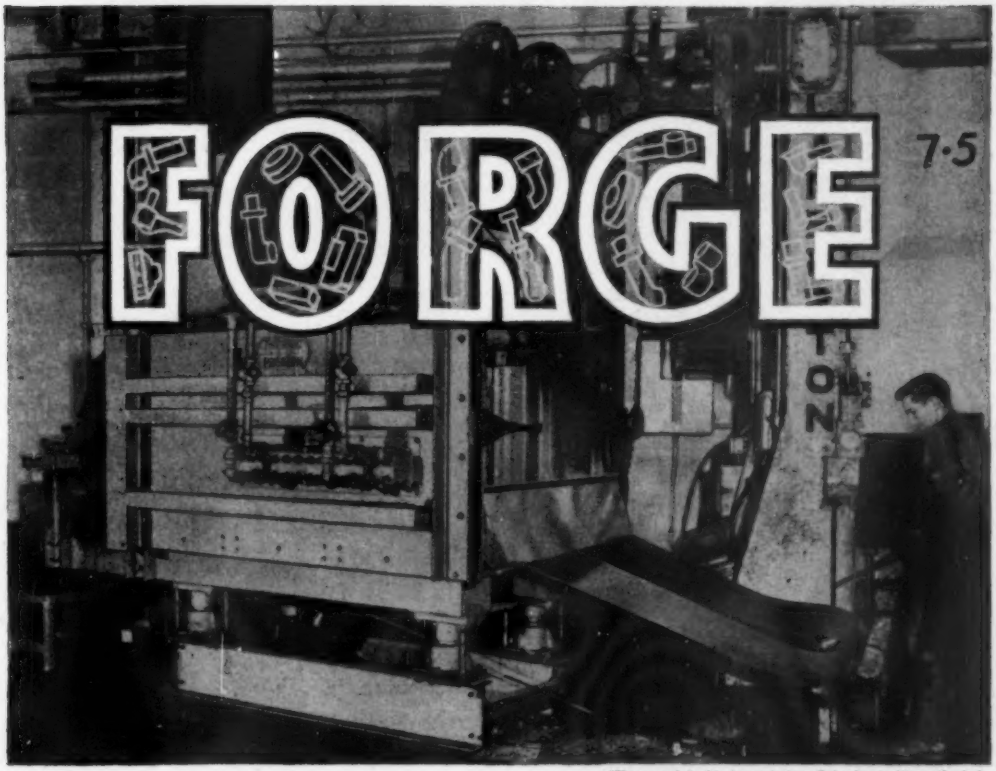

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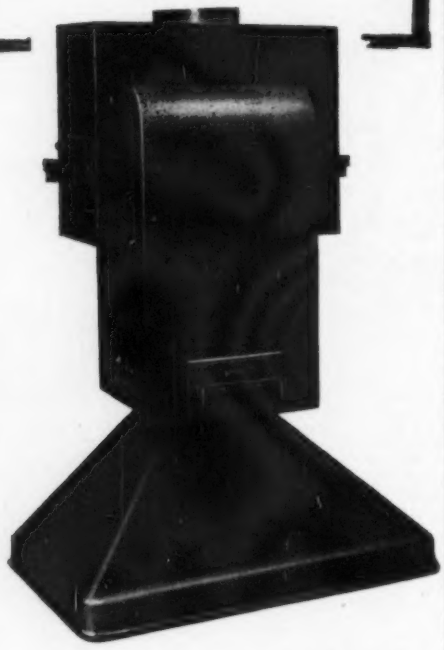

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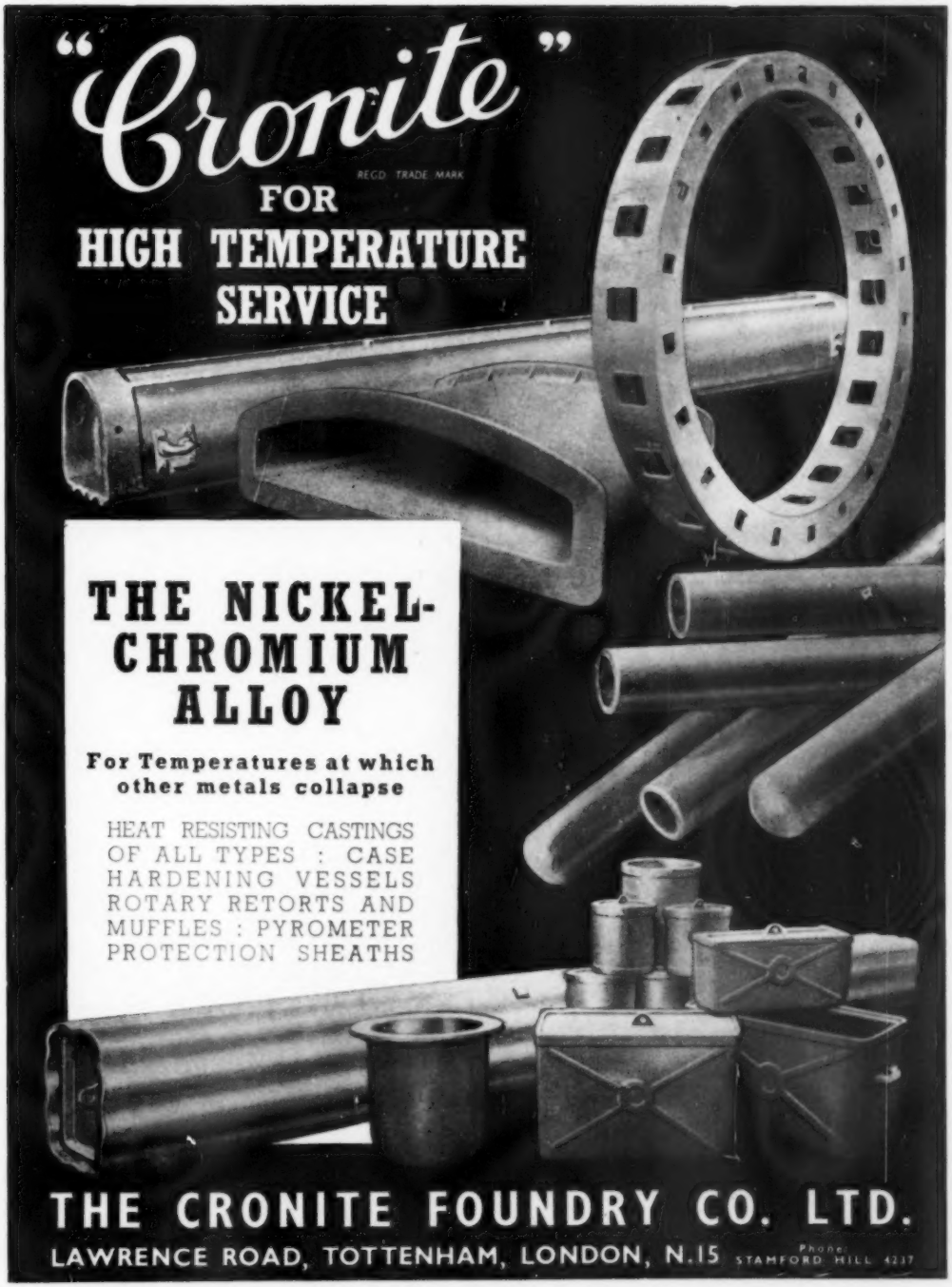
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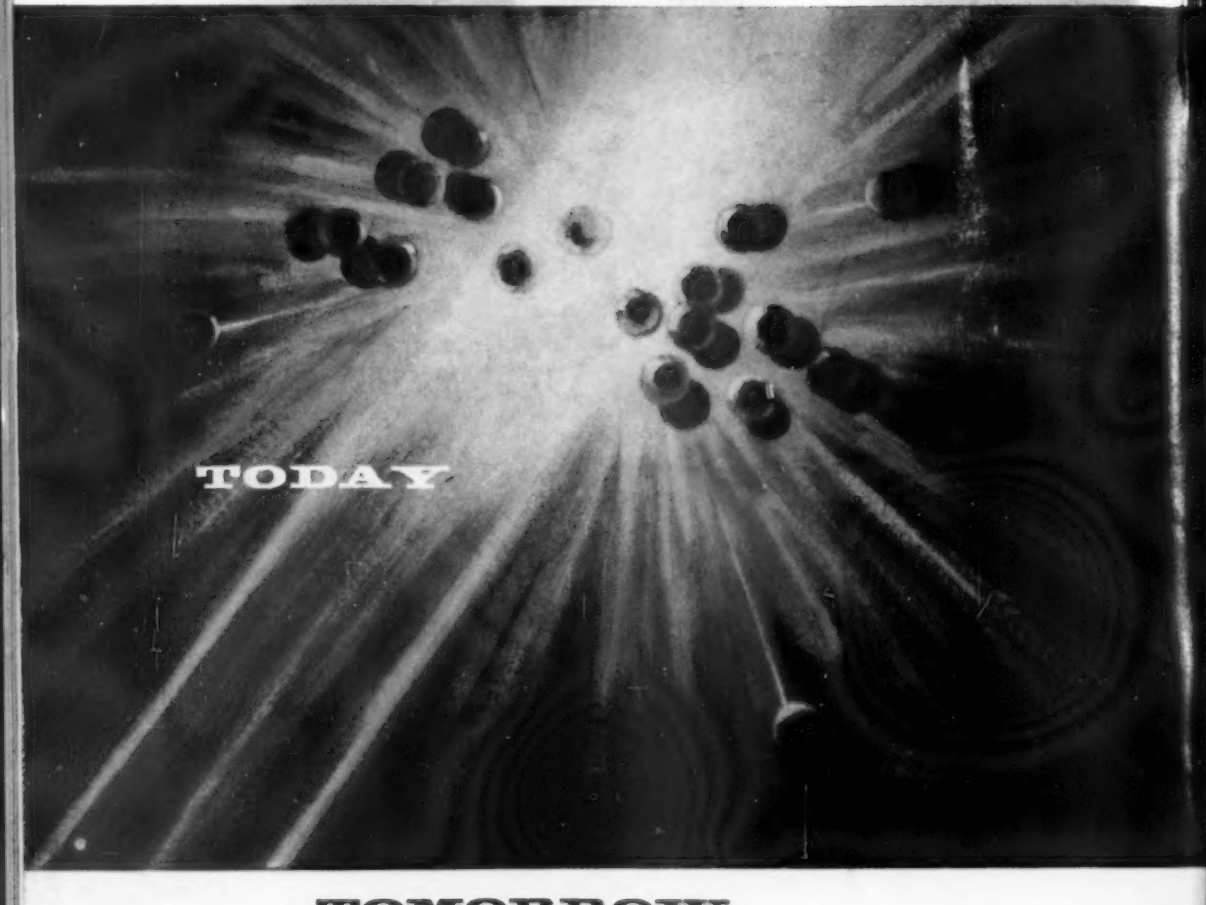

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